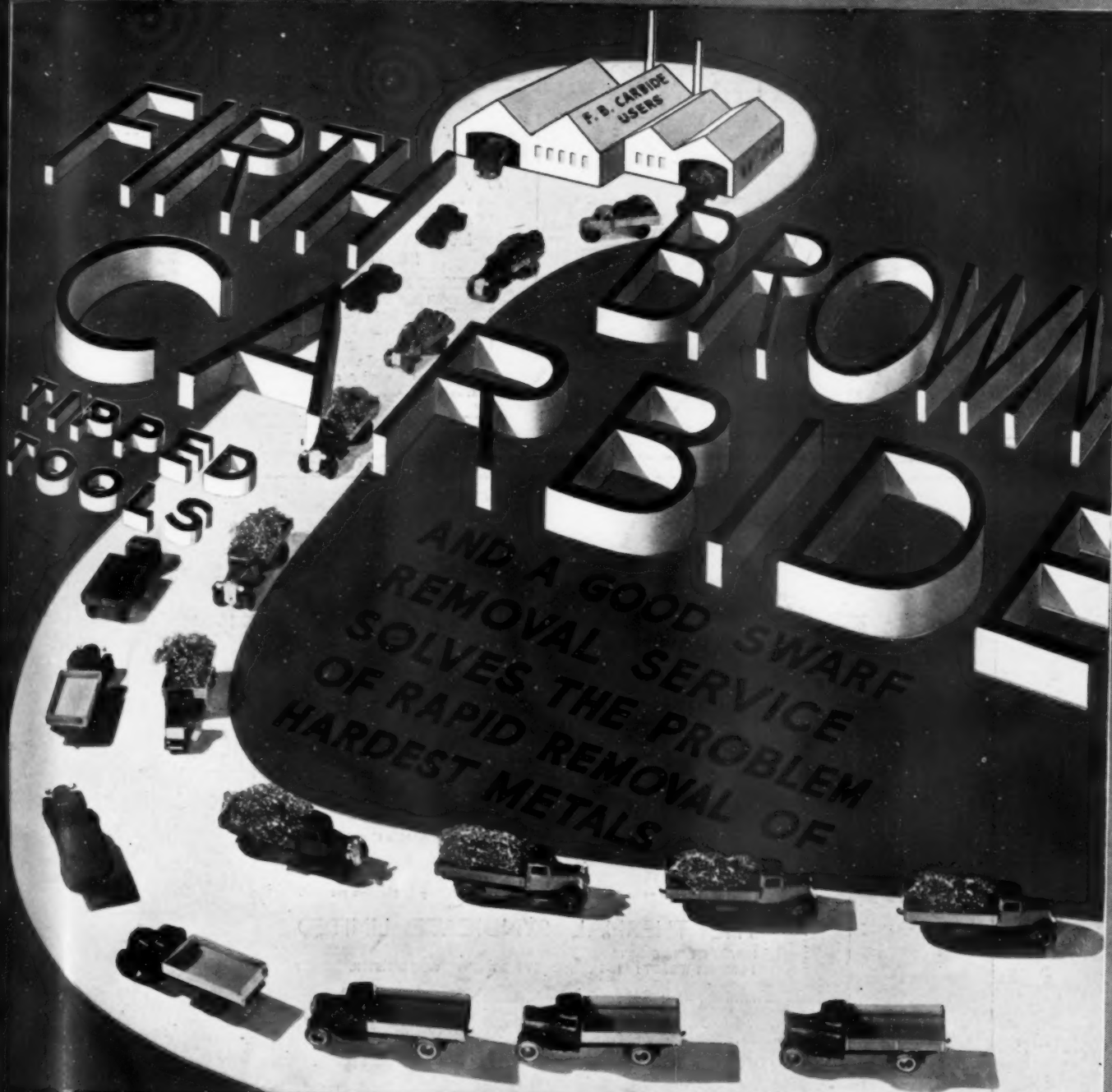


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THE BRITISH JOURNAL OF METALS

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METALLURGIA

The British Journal of Metals

(INCORPORATING THE METALLURGICAL ENGINEER.)

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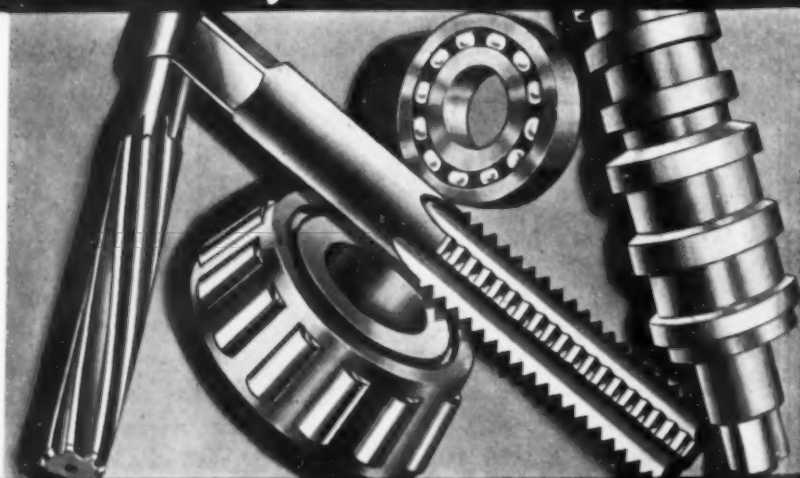
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METALLURGIA

THE BRITISH JOURNAL OF METALS.

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Mechanical Plating for Small Components

By E. E. Halls

The principal considerations governing the installation of mechanised plating plants are outlined, and reference made to the simplification and standardisation of the work to be done. The functions of the plant, its maintenance and construction, are indicated. The author stresses the fact that mechanised electroplating is not automatic plating, but that future development will be the merging of the two.

THE principles and practice of electroplating in batch-type equipment are well established, and most engineering factories operate plating shops which range from small to large, and which comprise a number of individual units from which the product is produced to meet a predetermined quality standard. The change over to a single mechanised plant which embodies all, or at least, the majority, of these individual units involves a large number of problems the solution of which has to be effectively achieved. Some of these can be foreseen and settled prior to the installation of the plant, but many of them can only be settled when the equipment is put into commission. Consequently, the elucidation of this latter group of difficulties has to be effected rapidly, and some foresight into the possibilities is therefore valuable. Mechanised electroplating plants are frequently loosely described as automatic equipments with the significance of the latter expression extended to all phases of the process. In point of fact, these equipments are in essence mechanically-operated devices for automatically transporting work through a given series of chemical or electrochemical immersion processes. Some of these with respect to certain features can be automatically controlled, but in the main they are manually operated. Consequently, with a clear appreciation of the true nature of these equipments to the fore, the problems which confront the engineer when changing from batch to continuous electroplating can be more truly realised. Conveyorisation of factory methods and processes in general have proved an immense success when properly worked out, especially in maintaining constant and uninterrupted flow of work, in the avoidance of pile-up of work around benches or machines, and in taking articles consecutively to the various operations that have to be imposed on them, as well as in the minimising of manual effort. These conveyorised systems are established and well known for automatically carrying work from department to department through the factory, for feeding individual machines and assembly bench lines, for handling work through metal finishing shops, and for specific processes such as heat-treatments and enamelling. In the case of electroplating, they have been in operation for extended periods in certain industries where the line of product does not involve a large number of component parts, and the disparity in size or shape so handled is not great. For the universal treatment of a miscellany of components, the electroplating problem becomes appreciably more complex. In the following, some aspects of this problem are discussed.

Dealing with the problem from the basic position of work arriving in the plating shop, articles of different

shapes and sizes, different weights, fluctuating combinations and quantities, different base metals, widely varying surface conditions with respect to contamination, corrosion or scale: all these have to be catered for. Additionally, the finish required may not be a single type of electroplating, but may involve several. Again, the plating specification to be complied with will not be the same for every article, different thicknesses of plate may be involved as well as mechanical operations of graining or polishing. This combination of diverse factors obviously has to be reduced to some simple formula which will render mechanised plating possible without excessive complications. It is necessary therefore to review the situation along the following lines.

- (1) Fix the number of electroplate finishes necessary.
- (2) Fix the number of finishes with respect to each electroplate finish and to thickness of deposit.
- (3) Fix the number of base-metal types.
- (4) Determine the limiting dimensions of the various articles to be handled.
- (5) Consider the surface condition of the work, and whether it shall be the duty of the conveyorised plating plant to cater for this condition, or whether work shall preliminarily be converted into a "universal clean condition."

Brief explanation of each of these is necessary.

Regarding (1), few engineering factories operate with only one electroplate finish, and most employ at least four. Obviously, some of these will be used to a very minor extent, and some to a major extent, and it is evident that only the latter group can be considered for mechanisation. It is necessary, therefore, to decide which these shall be, reduce the number to a minimum, and transpose as much work from the others as possible over to the electroplates so selected. The popular range of industrial electroplate finishes are nickel, copper-nickel, nickel-copper-nickel, nickel-chromium, zinc, cadmium, copper and tin. Given that all of these are involved, the most probable outcome would be the reduction of this imposing list to two platings. The nickel series would be relegated to the one for which greatest demand was experienced, and, taking advantage of development in recent years, the aim should be to adopt the direct nickel coating, or nickel chromium, according to local circumstances. Only one of zinc and cadmium would be mechanised, again the decision being made on the basis of the class of work and the market catered for. Copper and tin would probably be found employed only to a limited extent for special purposes, and would not be included in the schemes except insofar as they might be replaced to some extent by one of the

others selected. For clarity in presenting the following subject matter, it will be arbitrarily assumed that nickel-chromium and cadmium are selected as the two platings.

Concerning (2), mat, polished, and grained effects enter, and if any condition of the material surface other than its natural one is involved, it will enforce that surface conditioning noted under (5) must be catered for before the work goes to the mechanised plating plant, because after all cleaning is performed the graining or polishing must be effected. This, once determined, offers no further difficulty. Regarding thickness of coating, it may or may not be possible to compromise and evolve a single specification. The determining factors are usually (a) work for internal service, (b) work for external service or for tropical conditions, (c) impositions by virtue of design with special reference to assembly fits and operating conditions, (d) type of base metal. Decision to adopt a common thickness of coating simplifies plant and process operation, but is not essential.

Base-metal types, item (3), generally segregate themselves into aluminium-base alloys, zinc-base alloys, other non-ferrous metals comprising brass, phosphor bronze, copper, and copper-rich metals, and the ferrous metals consisting of iron and the carbon steels. The first two groups—aluminium and zinc—must each have a separate plant. The other two groups, non-ferrous and ferrous preferably, should have separate plants, but not necessarily so.

Item (4) assumes importance for several reasons. Firstly, the largest components in conjunction with output determine the size of the plant, and therefore largely influence initial expenditure, size of plant, floor space, etc. Secondly, the plant, for cost reasons, must be operated above a certain percentage of full capacity, as well as for control and efficiency reasons. It must be borne in mind that anodes are usually "fixed," and spaced to suit the general run of work. They cannot readily be withdrawn or changed at will as in small vat working, and consequently anode-cathode efficiency ratio will fluctuate widely with consequent troubles if the work is not controlled with some reasonable regard to total surface area. Usually, it is a question of deciding the range of work which can efficiently be handled; a small percentage of larger work must be excluded and treated elsewhere, likewise comparatively very small articles.

Surface condition, item (5), is largely a matter of local conditions. A study of the conditions in many factories has shown that bulk treatment to get the work into a reasonably clean condition before placing on the mechanised plating line is advantageous. With regard to aluminium and zinc-base articles, these will almost certainly be contaminated with cutting compounds and swarf, but otherwise unaffected. Degreasing by any suitable means, washing machine or trichlorethylene plant, leaves them in a condition readily suited to a mechanised line. The non-ferrous copper series will additionally be in a variable condition, some being tarnished, some having heat-treatment oxides or scales. In this case, a bulk degreasing and acid clean, either pickle or bright dip, is desirable. The ferrous metals will additionally have the possibility of rust, scale or heat-treatment films. A pre-cleaning is almost essential because of the varying times involved for pickling; also, sooty films may be present, and these can only be removed mechanically, by scratch brush or abrasive blasting.

These considerations narrow the proposition down along broad lines to the following:

- (1) Reduce to two finishes.
- (2) For each finish, a single thickness of coating can most probably be adopted; if not practicable, plant to be capable of control to provide two or more predetermined thicknesses.
- (3) The number of plants, determined by the number of finishes and the number of base-metal types.
- (4) The capacity of the plant.

- (5) Work cleaned to a suitable universal condition ready for the mechanised line.

With these questions decided, the handling of the work by means of wiring, racking, jigging or hooking, will one or all be adopted according to the nature of the articles. By these means, the work is attached to the conveyer, and from this point until completion of the plating cycle, when the work is detached from the conveyer, the plant must take care of all the processes involved in producing the required electroplate coating. To illustrate the requirements and functioning of the associated processes the cases of nickel-chromium on ferrous and non-ferrous metals, and of cadmium on the ferrous group will be briefly detailed.

For the nickel-chromium finish on both ferrous and non-ferrous (aluminium and zinc alloys excepted) metals, the following processes as represented by provision of tanks or vats are involved, the work naturally having been cleaned and prepared to a universal condition, viz., degreased, rust, scale and corrosion products removed, and appropriate surfaces ground and polished:

- (a) Electrolytic alkaline cleaner.
- (b) Water swill (hot).
- (c) Electrolytic alkaline cyanide etch.
- (d) Water swill (hot).
- (e) Dilute sulphuric acid dip.
- (f) Water swill (cold).
- (g) Bright nickel plating.
- (h) Water swill (cold).
- (i) Chromium plating.
- (j) Water swill (hot).
- (k) Water swill (hot).
- (l) Oven dry.

If brass, etc. is to be treated on a separate plant line from steel, the above sequence of processes can cater for the non-ferrous materials, but usually for the ferrous metals, it is preferred to substitute an anodic sulphuric acid etch (e), and to delete (c). Again, if racks or jigs carrying the work are immersed or partially immersed in the electrolyte, and are not rubber covered or otherwise protected from deposition, so that nickel-chromium is plated on to them, this has to be stripped off before re-use, otherwise the deposited metal flakes off and rapidly contaminates the plating solutions. The best procedure is, therefore, to incorporate two further processes at the beginning of the line, the first being an electrolytic caustic stripper and the second a water swill. Alternatively, manual stripping in a separate unit can be employed.

For the cadmium finish on ferrous metals, again assuming that pre-cleaning for grease, scale and rust has been exploited, a closely similar sequence of operations can be followed, the following being the simplest series:

- (a) Electrolytic alkaline cleaner.
- (b) Water swill (hot).
- (c) Water swill (hot).
- (d) Cadmium plating.
- (e) Water swill (hot).
- (f) Water swill (hot).
- (g) Oven dry.

In this case, jigs and racks may periodically be stripped down to the base metal in a separate unit.

Output and dimensions of individual articles having broadly determined the capacity of the plant, detailed sizes of the component units have to be derived from the immersion time required in each and from the maximum time allowable to carry over from one tank to the next. The plating operations proper are primary determinants since a specification for deposit thickness will almost certainly have to be worked to, so that length and speed of travel through the plating electrolytes will be important. Temperatures of solutions and swills have to be given attention, so that, in conjunction with the times allowed for traversing from one process to the next, drying off of the work (which is detrimental to adhesion) does not occur.

Control must be instituted to ensure that, having decided upon appropriate compositions for the various solutions,

these compositions are maintained within safe limits. The same applies to temperatures; the plating electrolytes would be provided with automatic means of control, and this is preferred, too, on water swills, although it is not essential providing the means for manual control and supervision are simple and efficient. Electrolytes also must be kept clean, free from sludge and dirt, and continuous filtration is essential on the main plating vats, nickel and cadmium in this instance; it is not usual to filter the chromium solution as immersion time is so short, and the degree of sedimentation in this is small, but, nevertheless, it is imperative that it should not be contaminated.

It will be appreciated that the plant will involve numerous pipe-lines, and many valves. Water lines for supplying or initial make up of tanks; steam lines for heating; air lines for agitation; and circulating systems for pumping, filtration, emptying, etc., all these must be maintained in good order, with special attention to valves, pumps, and steam traps and ensurance that no leakage occurs or failure of valves to cut off exists. Appropriate materials of construction must be selected. All plating tanks would be hard-rubber lined, except for chromium, which would be antimonial lead. All water swill tanks can be of welded steel construction lead lined, although the lining is only essential following sulphuric acid treatments. Air agitation coils immersed in the various plating solutions or swills are of ebonite. Busbars are copper, brass or aluminium. Pumps for circulating solutions are ebonite or glass lined. For filtration, there is a choice of plate filters, candle filters or metafiltration type of pack filter.

The mechanical features of the conveyers with respect to traversing and lifting devices are largely of a specialised nature, now in a highly developed stage by makers of these

plants. Nevertheless, it is important for the purchaser to satisfy himself that the plant is of such a flexibility that it accommodates the speed variations necessary to cater for different classes of work and different loads involved, and that it can do this without fouling anywhere or without special steps being taken with respect to jig sizes and spacing.

In a broad manner, the principal considerations governing the installation of mechanised plating plants have been outlined. Briefly, have been mentioned the considerations that the user must give with respect to the work to be handled through the plant with special respect to simplification and standardisation. Also, the functions of the plant, its maintenance and construction, have been indicated. Mechanised plating is a practicable proposition, and can be applied to most types of work, including the handling of articles of widely divergent classes through the same plant. Whether it is economically practicable depends upon the particular conditions obtaining, but only does it become a sound proposition if a rational interpretation and application of the analysis of the situation as outlined above ensures. The latter involves one further duty on the part of the user: the plant must be operated to full capacity and the load passed through as uniformly as possible with gradual gradation with respect to article size or type. This does not entail great difficulty, because the load for plating will be known at least eight hours ahead, and all that is involved is a system of segregating work into appropriate categories. Finally, it is again stressed that mechanised electroplating is not automatic plating, but it does not follow that the two are antagonistic; in fact, the future development will be the merging of the two.

Safeguarding Supplies of Strategic Minerals

THE outstanding feature of the use of metals in modern times is not the newness of these elements, but rather the ever accelerating quantity used and the ever-widening application of their uses. The growth of industry, particularly in the initial stages, was due to the close association of suitable minerals, and the close association of ore, fuel, and flux, not only made possible the improvement of machinery, but also increased facilities for the transportation of raw materials and their products. Coal was mined in small quantities in the eighteenth century, the amount obtained grew appreciably in the early decades of the nineteenth century; and since 1900, more coal has been extracted from the earth than in all the previous history of the world. The production of iron, copper, lead and zinc proceeded on similar lines.

As industry developed technique became more complex, more refined and exacting, and other metals with particularly distinctive properties were discovered and developed. Aluminium, although a relatively new metal, experienced a startling rise in world interest. The same applies to magnesium, though not to the same extent. These metals undoubtedly owe their rapid rise in application to the development of aeronautical engineering; but many previously obscure metals became indispensable for alloying steel. When Sir Robert Hadfield produced his special manganese steel in 1888, he indirectly opened the way for the production of other ferro-alloys, and extended the application of metals which had previously only been of interest in the laboratory. Such metals as chromium, cobalt, molybdenum, nickel, tungsten, vanadium, also gradually became associated with steel and their adoption opened up a new period of mineral development. It became necessary to look farther afield for supplies of these metals that had been regarded as rare in quantity

and therefore high in cost, this, together with the rise of industry, led to an unprecedented demand for metals and minerals, and to the discovery of increased sources of supply. But the exploration of geologists and mining engineers shows that the essential mineral products are far from being evenly distributed over the land areas of the world.

With the increasing number of metals applied in industry national sufficiency in mineral raw materials, if it ever existed, is unknown to-day. In normal peace times nations are concerned with the problem of obtaining adequate supplies for an indefinite future, and are more immediately concerned when a mineral resource, however abundant its total world supply may be, is not found within the boundaries of the nation needing it. In such cases international arrangements are usually made to safeguard, as far as possible, the uninterrupted flow of the minerals to the consuming nation, and ships are sent to all quarters of the globe to gather them.

It will be appreciated that nature has not taken into account the political boundaries of nations in its distribution of mineral supplies. Some minerals are so widespread that nearly all countries have adequate supplies within their own boundaries, or, at least, near at hand. Other minerals are so distributed that some parts of the world have a surplus and others a deficiency. No country is entirely self-sufficing in regard to either supplies or markets for all mineral commodities, so that international exchange of minerals cannot be avoided if all nations are to be supplied with needed materials. In peace times the necessary international movement of minerals may be aided or hindered by various artificial means, but the international channels of mineral movement are not altered by legislation. When a major war is in progress, however, the situation is radically

changed, and supplies of materials may cease to be available to some countries through difficulties of transport.

Europe, as a whole, apart from Russia, is not very well endowed with mineral resources, and though some countries are better off than others—Spain, for instance—no single one, and indeed, no contiguous group, is self-sufficing. And, yet, within the British Empire itself there are sufficient metals and minerals not only for its own requirements, but, in most cases, to provide a surplus for use by other countries against which she is not at war; further, she is in a position to transport her supplies from overseas. With the progress of the defence programme of the United States a strategic minerals survey is being conducted jointly by the Bureau of Mines and the Geological Survey.

The United States is one of the countries richest in mineral deposits, but it is far from being self-sufficient in strategic minerals. However, the general outlook for locating important deposits in that country is definitely more encouraging than was anticipated at the beginning of the search. According to Dr. R. R. Sayers, acting-director of the Bureau of Mines, the situation with regard to five minerals—essential in peace and critical in war—is somewhat promising. These minerals are manganese, chromium, tungsten, mercury and antimony. The nickel situation is less encouraging and results in tin are classed as rather negative.

The importance of locating and developing useful domestic deposits of the major strategic minerals is shown by the heavy percentage of these minerals which must be imported at the present time. The following table, according to the Bureau of Mines, shows the percentage of United States peace-time requirements produced from domestic mines during a recent five-year period:

| | |
|-------------------|------|
| Manganese | 5-6% |
| Chromium | 1% |
| Mercury | 40% |
| Tungsten | 50% |
| Nickel | 0.5% |
| Tin | 0.2% |
| Antimony | 10% |

Actual exploratory work has been conducted by the Bureau on nine different projects, of which one is an antimony deposit in Idaho, three are chromite deposits in Montana, Wyoming and Oregon, one is manganese in Washington, one nickel in Nevada, two tin in South Dakota and New Mexico, and one tungsten in Nevada. The results have been sufficiently encouraging to warrant further work on six of these projects, and on four of them this further work is now being carried out, or is being planned. In all of the projects examined some new ore bodies were found, though they were not generally of commercial grade. It is pointed out, however, that, for availability as strategic reserves, it is not essential that deposits be of commercial grade.

On manganese, the most important of the strategic minerals, mainly because of its vital role in the making of steel, actual exploratory work was done only in one locality, for the reason that there were so many manganese deposits to be considered that the most promising ones could not be selected immediately. It has now been possible to select from the manganese deposits considered 47 that warrant exploratory work, and during the next fiscal year it is proposed to give precedence to the exploring of manganese deposits. The United States reserves of low-grade manganese deposits are large; the problem being the devising of metallurgical processes that may make some of these low-grade reserves available for strategic demand. One small lens of high-grade manganese ore was discovered which will not add to the strategic reserves, because it is of such grade and quality that it probably will be quickly mined out, and the ore sold at a profit. Of interest, however, is the fact that the drilling cost was only about \$1 per ton of ore indicated, the value of which is \$3 per ton or more at prevailing prices. Imports of manganese

ore in the past have been mainly from Russia, Cuba, the African Gold Coast, and Brazil.

Extensive deposits of chromite of a grade that could probably be used in an emergency are indicated in one locality. Further exploratory work is in progress for the purpose of definitely proving the existence of sufficient tonnage to be of real strategic importance. Metallurgical investigations on this ore, which is not up to standard grade, indicate that a high-grade ferro-chrome product can be made, though at a cost somewhat above the normal price of this alloy. Imports have been largely from Southern Rhodesia, Union of South Africa, Philippine Islands, Turkey and New Caledonia.

A large low-grade deposit of antimony ore has been indicated by diamond drilling. Although sub-commercial in grade, this deposit could furnish a substantial percentage of United States requirements in an emergency at a price considerably above the prevailing one. Some high-grade antimony ore has been discovered occurring in small lenses, and it is believed that, in the general area in which they occur, there are enough of these small deposits to constitute an important reserve. The greater part of United States antimony imports prior to 1931 came from China; since that date Mexico has become the principal source of imports.

The outlook for the United States becoming self-sufficient in tungsten is very good, largely as the result of the efforts of private enterprise.

In the case of mercury, the prevailing high price has stimulated production from lower-grade domestic deposits during the last few months, to a production rate equal to the requirements of industry, although the length of time this rate of production can be continued is problematical. Mercury is used in the manufacture of fulminate, for detonating high explosives and fixed ammunition, for gold recovery by amalgamation, in drugs, dental amalgam, and for many other purposes.

Two or three nickel-bearing deposits were discovered that are of sufficient promise to warrant exploratory work and sampling. The great bulk of production is from Canadian deposits.

As in the past, tin reserves have not been evidenced. It would be impossible to produce any large amount of tin, even from the low-grade ores found. One small ore shoot was partially delineated to a depth of about 300 ft. and found to contain enough tin that it could be worked at prices about double the prevailing market price. There might be a considerable amount of similar material at depths below that which was prospected. The United States normally consumes about 75,000 tons of metallic tin or half of the world's supply. In recent years, about 75% of the world's supply of tin has been derived from placer deposits in the Malay States and the Dutch East Indies.

There is great activity in the Canadian mining industry, and the production of such metals as copper, nickel, zinc antimony and cadmium is reaching new high levels. In this country, too, considerable prospecting and development work has been carried out in connection with the search for metals and ores which have not as yet been produced to any great extent in Canada, but which are strategic for war purposes in the manufacture of various alloys. These include such minerals as molybdenite, manganese, mercury and tungsten. Canada is exporting large supplies of copper, lead, zinc, nickel and aluminium to Great Britain.

All industrial countries, whether at war or not, find it necessary either to carry out exploratory work for new ore bodies, or to conserve the minerals they possess to ensure adequate supplies, but it will be appreciated that the accelerated rate of mineral consumption due to war conditions is causing almost all nations to view the question of supplies with some misgiving, and forcing them to take steps to safeguard as far as possible, supplies of the major strategic minerals.

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Secondary and Scrap Aluminium

WITH the ever-expanding metal requirements, the need for the conservation of metals has become very real in times of peace; common sense as well as economic necessity has indicated the more prudent use of the ores available, without waste or needless destruction. Conservation is, of course, more essential to a country at war when waste of metals cannot be avoided, and while attention must be given to safeguarding supplies of strategic minerals, it must be remembered that the stock of metal in use is no less a valuable resource than unmined ore.

It is frequently assumed that conservation is concerned only with the production phases of the metal industries, whereas the effective conservation in the utilisation phases is, at least, as important. It should not be overlooked that real conservation can be achieved by promoting the constant use and reuse of that part of production which is not destroyed in service. Thus, the recovery of scrap and wastes, and their use with or instead of metals produced from ores, is sound economics, and entirely in accordance with the general principle of conserving natural resources. This principle is recognised in all the metal-producing industries. In the iron and steel industries, for instance, the use of scrap has become essential in the manufacturing technique of steel, but for some unaccountable reason efforts have been made in this country to prevent the re-use of aluminium.

A few weeks ago a great Press campaign was launched, at the instigation of the Minister for Aircraft Production, for scrap aluminium. The nation was led to believe that the country was in urgent need of this scrap, and an appeal was made for all kinds of household utensils fabricated of this metal. Actually, there is no shortage of bauxite in the country and ample supplies of primary aluminium are available, but the remarkable fact is that many thousands of tons of secondary aluminium were also available at the time the appeal was made, yet permission for its release for use in the aircraft industries, for engineering purposes connected with export, or even for home consumption, could not be obtained.

The response to the appeal was immediate and resulted in a very considerable quantity of scrap aluminium becoming available, but it has caused some embarrassment, since large quantities were already tied up in the hands of merchants and melters and the possibility of its release by the controller for work of national importance seemed to be remote. We understand that some action is to be taken to release this valuable resource, and there should be no delay in the conversion of this scrap to secondary metal, so that it can be used on essential production.

The term secondary metal is used to distinguish the product from virgin metal, and this distinction does not imply that secondary metals are of inferior quality, because metals derived either from ore or from scrap vary in purity and adaptability to use in making certain products. It is true that the specifications for certain products stipulate the use of virgin metal, but, providing reasonable care is taken in the selection of scrap for conversion, secondary metal can be used successfully for about 75% of the aluminium alloy products. The surprising fact is that not only has this valuable resource been debarred from use for

aircraft production, but its release has been refused for products intended for export and home industries not connected with war services.

Much has been said for and against the use of secondary and of virgin metal, and there is no doubt that when producers concerned in the recovery of metal from scrap first commenced operations, there was ample reason for criticism, but at that time they had no adequate technical control of operations, whereas to-day technical advancement has facilitated the production of reliable secondary metals, and the progress made in their use under peace conditions contributed considerably in conserving precious supplies of the world's metals. Immediately prior to the war secondary aluminium was a recognised source of supply for a wide range of products, and while it can be readily understood that control of such a vital metal was necessary to ensure that available supplies should be directed into proper channels, it does not surely mean that these supplies should be held up indefinitely.

It is true that the presence of contaminating substances in scrap aluminium has considerable metallurgical importance apart from the harmful effect of such substances in lowering the value or rendering the scrap more troublesome to treat and utilise in the production of secondary metal. In addition to metallic impurities in scrap, which may be alloyed with the metal or mechanically mixed, various non-metallic impurities are present. The scrap aluminium may include dirt, paint, oil, water, silica, aluminium oxide, wood and other organic substances, and the secondary metal produced from such scrap may contain aluminium oxide, aluminium carbide, aluminium nitride, and such dissolved gases as hydrogen, carbon dioxide, and hydrocarbon gases. But many of the problems of contamination of aluminium scrap are solved by proper sorting and grading, and much can be done by suitable treatment in the remelting operation to make the secondary metal of good quality. If clean scrap is used and it is carefully graded and remelted under controlled conditions, there is no reason why the secondary metal produced should not compare favourably with certain grades of virgin metal.

Accurate information regarding the quantity of aluminium in the form of scrap or remelted and available as secondary aluminium is not known, but the amount held in this country due to the methods of control adopted must reach very high figures. The majority of this is in the hands of remelters, who are specially equipped and organised to market the material in suitable forms for the consuming industries. In normal peace times it is probably safe to say that of the aluminium put into industry in the course of a year about one-third consists of secondary aluminium and scrap. Thus, for every pound of virgin aluminium produced about half a pound of secondary metal or scrap is used. It is difficult to appreciate why present conditions should cause restrictions to be imposed which almost exclude the use of secondary metal and scrap. In fact, there is some excuse for believing that the amount used in production would be greater rather than less, always providing that stocks were available.

The recovery of aluminium from scrap of various kinds and its use with or instead of virgin metal is sound economics, certainly the existence of a large floating supply of scrap should not be ignored. It should be remembered that the bulk of aluminium produced is still in existence in

structures or as finished manufactures. Apart from the aluminium used in the manufacture of steel, for thermit metallurgy, paint manufacture, and several other purposes of relatively minor importance, most of the existing metal will return to scrap and reappear on the market in a suitable form to re-enter production, and it is an economic sin to use virgin metal when this secondary aluminium or scrap will serve equally well. Not only is the use of secondary aluminium and scrap necessary to conserve natural resources, but it is less expensive and therefore more economical to the consumer. But many consumers do not yet like the idea of using secondary metal with or instead of virgin metal; the term seems to indicate a second preference that implies a lower quality. In recent years, however, there have been important developments in methods of production, and as a result there has been a great improvement in the quality and uniformity of the secondary metal available; the objections to its use, therefore, are not now so impressive as in former days. The fact that secondary aluminium is available at a lower price than virgin aluminium is not an indication of inferior quality; obviously, remelters must use the best possible methods in order that the secondary metal they market will compete satisfactorily with virgin metal. The shortage of virgin aluminium during the war of 1914-18 proved conclusively that secondary and scrap metal could be used, and there seems no reason why its use should be withheld until there is again a shortage of virgin metal.

Minerals used in Welding

THE minerals that enter welding rod coatings and fluxes are the main subject of a comprehensive report by Ralston and Bernewitz.* Few mineral producers know just what minerals and what grades and quantities of minerals are required by the welding industry, and it is for their information that this report has been prepared. The various methods of welding are discussed, but welding experts will find of special interest the information on minerals and mineral products that they use.

Electrodes are available for all types of welding, including different metals. They may be either of metal or carbon, depending on the job. In manual welding, the metallic electrode generally is used. They range in size from $\frac{1}{16}$ in. to $\frac{3}{8}$ in. in diameter, and 14 in. to 18 in. in length. The proper size also depends on the class of work to be done. "Washed electrodes" are coated lightly with an arc-stabilising chemical such as lime. "Semi-coated electrodes" have a coating which generally is applied by dipping, and which constitutes 1 to 2% of the total weight of the electrode. Such a coating stabilises the arc and may partly control oxidation of the molten metal, as it is deposited by forming a thin film of slag over the surface of the bead. "Heavily coated shielded-arc-type electrodes" utilise all the benefits of chemical coatings. These may constitute a tenth or more of the total weight of the electrode, and they control the arc characteristics and the physical and chemical properties of the deposited metal. "Carbon electrodes" for manual welding and cutting are made in sizes $\frac{3}{32}$ in. to 1 in. in diameter and 12 in. in length. For automatic welding with the shielded arc, the electrodes are $\frac{3}{16}$ in. to $\frac{1}{2}$ in. size. Carbon-arc welding differs considerably from metallic-arc welding in that the heat is used only to melt or fuse the base metal. Filler metal is not commonly used.

Minerals used as stabilisers include titanium oxide (this enables a longer arc to be maintained), iron oxides, manganese oxide, lime, sodium and potassium minerals or compounds, and thorium oxide. They lessen wandering of the arc at the tip of the electrode. The stability of an arc of given length is the ratio of the distance the arc can be lengthened before being extinguished to the given arc length. Stability often includes smoothness and minimum spatter. Voltage and other factors affect arc stability. Steel deoxidised with aluminium is bad for a steady arc. Alkali compounds, especially those containing sodium and

potassium, have a marked stabilising effect on the arc, but the former accelerates and the latter retards the flow of metal. Calcium compounds steady the arc, as does iron oxide. Iron silicates vary in effect.

The subject of electrode fluxes is extremely complex, and involves a knowledge of metallurgy, chemistry, ceramics, and circuit characteristics.

Flux compositions are usually guarded with considerable secrecy. Fundamentally, they consist of acid, basic and/or neutral slag-forming materials, arc stabilisers, de-oxidisers, and/or alloying materials, organic material if desired, and a solution of soluble silicate for a binder. No sharp distinction can be drawn as to the function of any single ingredient of a flux, for each material directly or indirectly influences one or more of the functions of the composite flux. Silica, manganese dioxide, iron oxide, amphibole, feldspar, and titanium oxide or rutile may be considered as slag-forming materials, while ferro-manganese, ferro-silicon, and silico-manganese usually function in a dual capacity as deoxidiser and for alloying additions. Sodium silicate is almost universally used as a binder.

Coatings are really fluxes or reaction substances added to the outside of rods or electrodes to stabilise and protect the arc and to provide a method of depositing fluxes on the welds and form slag, which is easier than dipping bare rods into a container of flux and thus carry it to the point of welding. Coatings affect the ease with which the arc may be maintained and the character of the deposited metal. They modify the rate of fusion and depth of penetration. Also, certain treatment before application of the coating material facilitates adhesion of the coating and improves its effects. Coatings are discussed extensively under many headings, which include the manufacture of coated electrodes and information on particular minerals used; attention is also given to proprietary coatings. The report also includes considerable information on fluxes; slags; functions of constituents of coatings, fluxes, and slags; minerals and compounds used in welding; metals that can be welded and the common minerals used; manufacturers of welding rods and fluxes, etc.

Fusible Alloys in Toolshop and Foundry

UNTIL a few years ago the chief applications of fusible metals were for electrical fuses, fire alarms and sprinklers, temperature indicators, low-melting-point solders and similar purposes, but in the last few years they have acquired great importance in the tool-shop and foundry. Some fusible alloys developed for these purposes are discussed in the current issue of *Tin and Its Uses**. Three proprietary metals are referred to, marketed as "Cerromatrix," "Cerrobaze," and "Cerroband." These metals are stated to be accurately compounded alloys, Cerromatrix having been developed for the mounting of dies and punches in press-tool work; Cerrobaze, for the casting of models and patterns for foundry work and the production of moulds by electro-deposition; and Cerrobend for the bending of tubes and rolled or extruded sections.

All the alloys contain a preponderating proportion of bismuth which conduces to their low-melting points, and also causes them to expand slightly on solidification, a property that in each application is of the highest importance. Setting dies and punches in their holders by simply pouring a fusible alloy round them is not new, but lead, solder and other alloys, which do not expand on solidification, do not give a firm enough grip on the tools, and Wood's metal, although it expands, is not mechanically strong enough. It is claimed for Cerromatrix that its net expansion is about 0.002 in. per inch, while it is amply strong enough for the purpose. The method of mounting dies and punches is discussed, and useful information is given regarding the pattern metal and the metal to assist tube and section bending.

* Bureau of Mines Information Circular 7,197, U.S. Depart. of the Interior.

*Issued free by the International Tin Research and Development Council, Fraser Road, Greenford, Middlesex.

The Thermal Conductivity of Ferrous and Non-Ferrous Alloys

By J. W. Donaldson, D.Sc.

Recent investigations on the heat conductivity of many industrial alloys are reviewed and the resulting data given. The review includes work on carbon and alloy steels, grey and white cast irons, high-duty and alloy cast irons, copper and nickel alloys, and two papers dealing with the thermal and electrical conductivities of iron and steel are discussed.

SINCE the last review in METALLURGIA¹ of the work which had been carried out on the thermal conductivity of ferrous and non-ferrous alloys, a number of systematic investigations have been made with a view to determining reliable data on many industrial alloys in relation to their heat-conducting property. These investigations have included careful determinations on technically pure iron and on some carbon and alloy steels, data on a series of grey cast irons, determinations on high-duty and alloy cast irons, and determinations on some industrial alloys of copper and of nickel. The work on the series of grey cast irons also included determination of the electrical resistivities of the various irons. A survey has also been made of existing data on the thermal and electrical conductivities of irons and steels and fresh data has been determined on carbon and alloy steels and included in the Second Report of the Alloy Steels Research Committee of the Iron and Steel Institute.

Carbon and Alloy Steels

The researches on the pure iron and on the carbon and alloy steels were carried out by Esser, Eilender, and Pütz² in a specially designed apparatus on wire rod specimens 5 mm. in diameter and 2 mm. long, to which ten thermocouples were fixed at varying intervals. Measurements were taken, first by determining the rate of cooling of the uniformly heated bar, and then by determining the temperature gradient along the same bar heated locally at one end. To determine the thermal conductivity from these measurements, the specific heat and specific gravity of the various materials had also to be determined. In Table I are given the composition of the various steels, together with their thermal conductivities at 100°, 200°, 300°, 400°, and 500° C.

Consideration of these results shows that the values obtained for the electrolytic iron and for the low carbon steel are considerably higher than those obtained by previous investigators. In recent work on similar materials there has been a tendency to obtain increased values for such materials, and this may result from the purity of the specimens tested, and would indicate that a very slight increase in purity produces a marked increase in thermal conductivity.

The thermal conductivity of the six carbon steels shows a decrease as the carbon content of the steel increases and is in accordance with work done in previous investigations. At each temperature the decrease in conductivity is at first rapid, but becomes more gradual as the carbon content of the steels increases. A more pronounced fall in conductivity occurs, however, over the range 100° to 300° C. than that obtained over a similar range by other workers. The values obtained for the low alloy steels are in agreement with previous work and show a considerable decrease in conductivity due to alloy additions when compared with the carbon steels. The fall in conductivity with temperature is also much less than that obtained with the carbon steels, with the result that the values obtained for the alloy steels and the low and medium carbon steels at 500° C. are somewhat similar. The high alloy, austenitic steels

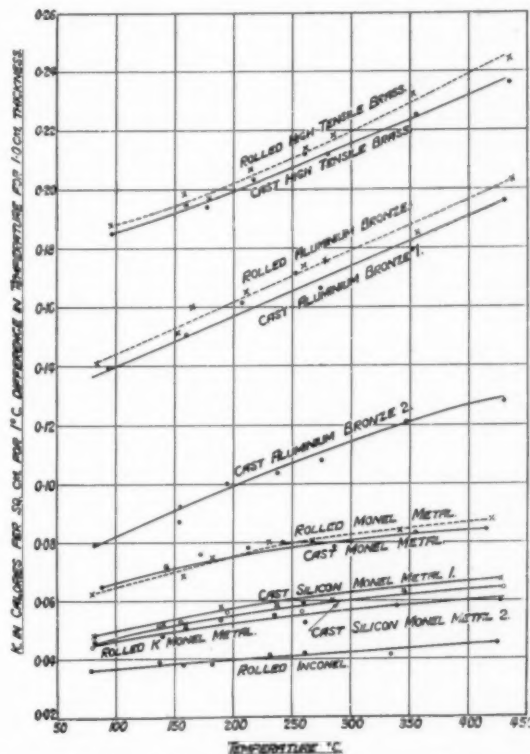


Fig. 1.—Thermal conductivities of copper alloys and of nickel alloy (Donaldson).

show a slight increase in conductivity with increase in temperature, but not so marked as that obtained in some previous investigations.

Grey and White Cast Irons

The most recent determinations made on the thermal conductivity of ordinary grey and white cast irons are those carried out by Maréchal and Listray.³ The electrical resistivity of the irons were also determined by the usual method. The thermal conductivities were measured in a specially designed apparatus over the range 150° to 200° C. The cast irons examined consisted of a series of grey irons, containing large, well-defined graphite flakes, and having an approximate composition of 3.0% total carbon, 2% silicon, 0.45% manganese, 0.66 to 1.56% phosphorus, 0.10% sulphur, and 0.15 to 1.36% copper, and a series of white irons containing very finely divided graphite, not at all uniformly dispersed and having an approximate composition of 2.7 to 2.9% carbon, 2% silicon, 1.2% manganese, 0.06 to 0.1% sulphur, and 0.46 to 1.96% phosphorus. Both series of irons were cast in bars 35 mm. and 75 mm. in diameter.

The thermal conductivity of the grey irons decreased with increase in temperature and with increase in phosphorus. The presence of copper in some of the irons had

¹ J. W. Donaldson, METALLURGIA, 1936, 13, 159-160.

² E. Esser, W. Eilender, and E. Pütz, Archiv. Eisenhüttenwesen, 1938, 11, 619-622.

³ J. Maréchal and J. Listray, Revue de Métallurgie Memoires, 1939, 36, 240-250.

TABLE I. THERMAL CONDUCTIVITY OF IRONS AND STEELS (ESSER, EILENDER, AND PÜTZ).

| Material. | Composition. | | | | K in Cal./Cm. Sec. Deg. | | | | |
|-------------------------|--------------|-------|------|------|-------------------------|---------|---------|---------|---------|
| | C. | Mn. | Ni | Cr. | 100° C. | 200° C. | 300° C. | 400° C. | 500° C. |
| Electrolytic iron | — | — | — | — | 0.208 | 0.184 | 0.157 | 0.134 | 0.120 |
| Carbon steel | 0.065 | 0.40 | — | — | 0.193 | 0.165 | 0.140 | 0.123 | 0.109 |
| " " | 0.29 | 0.84 | — | — | 0.180 | 0.154 | 0.125 | 0.105 | 0.0907 |
| " " | 0.52 | 0.63 | — | — | 0.162 | 0.132 | 0.109 | 0.0852 | 0.0750 |
| " " | 0.85 | 0.65 | — | — | 0.160 | 0.122 | 0.102 | 0.0875 | 0.0695 |
| " " | 1.10 | 0.55 | — | — | 0.156 | 0.119 | 0.100 | 0.0800 | 0.0650 |
| " " | 1.40 | 0.53 | — | — | 0.152 | 0.119 | 0.0961 | 0.0760 | 0.0601 |
| Ni-Cr Steel | 0.13 | 0.40 | 4.5 | 1.1 | 0.137 | 0.134 | 0.130 | 0.128 | 0.125 |
| " " | 0.30 | 0.60 | 1.5 | 0.5 | — | 0.073 | 0.063 | 0.060 | 0.058 |
| " " | 0.35 | 0.60 | 4.5 | 1.3 | 0.109 | 0.101 | 0.090 | 0.080 | 0.071 |
| Austenitic steel | 0.12 | — | 18.0 | 27.0 | 0.049 | 0.050 | 0.052 | 0.053 | 0.054 |
| " " | 1.12 | 13.50 | — | — | 0.0442 | 0.0448 | 0.0455 | 0.0478 | — |

no marked influence. The influence of phosphorus is similar to that obtained by the present author⁴ in previous experiments, when it was found that an increase of phosphorus from 0.18 to 0.95% decreased the conductivity from 0.115 to 0.110 in an iron containing approximately 3.3% carbon, 2% silicon, 0.76 to 1.0% manganese, and 0.06% sulphur. The thermal conductivity of the 0.75 mm. diameter bars was greater than that of the 0.35 mm. diameter bars of the same iron. The values obtained for the white cast iron were considerably less than those obtained for the grey irons, but the material of these irons was so inhomogeneous that it was difficult to draw any definite conclusions as regards the influence of phosphorus on the heat-conducting or electrical resistance properties.

With regard to the electrical conductivities, those of the cast 75 mm. diameter bars were smaller than those of the 35 mm. diameter bars, and the electrical resistance of the latter bars increased with increase in phosphorus content, whereas the composition of the former bars had no distinct influence. The electrical conductivities of the white cast irons were distinctly greater than those of the grey cast irons. The results also tend to show that the Wiedemann-Franz-Lorenz law does not apply to cast iron.

High Duty and Alloy Cast Irons

During recent years considerable advances have taken place in the metallurgy of cast iron and have resulted in the development of the high duty and alloy cast irons. As it has been found that alloy additions exert an important influence on the heat-conducting properties not only of steels but also of cast irons, it was considered of interest to investigate this property in those newly developed irons with a view not only of determining the influence of the combination of alloying elements already experimented with, but also of determining the influence of new alloying elements. With this aim in view, a series of investigations were carried out recently by the author.⁵

Three series of cast irons were tested over a range of temperature from 40° to 530° C. The first series of four irons consisted of an unalloyed iron and an iron of similar composition with the addition of 1.58% of copper, a second unalloyed iron and another of similar composition with 0.54% of chromium and 0.77% of molybdenum added. The irons in the second series were high-duty iron with tensile strengths of over 22 tons per sq. in., and included an unalloyed low-carbon iron, a molybdenum iron, a Ni-Tensyl iron, a manganese-nickel iron, and a nickel-chromium iron. The third series consisted of the heat-resisting irons Ni-Resist and Nicrosilal, and an aluminium-chromium iron containing 7.0% aluminium, 1% chromium. The composition of those various irons and their thermal conductivities at 100° and 400° C. are given in Table II.

Previous investigations⁶ have shown that 0.8% of silicon lowers the thermal conductivity of grey cast iron by approximately 0.01, and that 0.75% of nickel produces a fall in conductivity of 0.013. The influence of copper appears, therefore, to be similar to that of silicon and nickel, but is not so marked. From the values obtained with the copper cast iron when compared with the unalloyed iron of similar composition, it appears to be about half that of silicon, whereas nickel is probably about 30% more than that of silicon. The influence of copper in lowering the conductivity is not in agreement with the results obtained by Maréchal and Listray,³ who found that copper had no marked influence on the thermal conductivity.

Molybdenum appears to raise the thermal conductivity of cast iron in a similar manner to chromium and tungsten, but its influence is not so pronounced as in the case of these elements. Previous investigations have shown that a rise in the thermal conductivity of approximately 7.5% is produced by 0.4% of chromium or 0.5% of tungsten. To produce the same increase with molybdenum, a content of 0.6% is necessary. The values obtained for the high-duty irons, Ni-Tensyl, manganese-nickel, and nickel-chromium show the influence which the special element additions in

4 J. W. Donaldson. *J. Iron and Steel Inst.*, 1933, 78, 255-276.

5 J. W. Donaldson. *Proc. Inst. British Foundrymen*, 1938-39, pp. 125-132.

6 J. W. Donaldson. *Proc. Inst. Mech. Eng.*, 1928, 4, 953-983.

TABLE II. THERMAL CONDUCTIVITIES OF PLAIN, HIGH-DUTY, AND ALLOY CAST IRONS (DONALDSON).

| Cast Iron. | Composition. | | | | | | | | K. | |
|--------------------------|--------------|------|------|-------|------|------|------|------|---------|---------|
| | T.C. | Si. | Mn. | Ni. | Cr. | Mo. | Cu. | Al. | 100° C. | 400° C. |
| Plain | 3.20 | 1.56 | 0.72 | — | — | — | — | — | 0.121 | 0.108 |
| Copper | 3.18 | 1.58 | 0.69 | — | — | — | 1.58 | — | 0.112 | 0.101 |
| Plain | 3.11 | 2.26 | 0.39 | — | — | — | — | — | 0.111 | 0.101 |
| Copper-molybdenum | 3.12 | 2.31 | 0.39 | — | 0.54 | 0.77 | — | — | 0.119 | 0.109 |
| Plain | 2.61 | 2.46 | 0.45 | — | — | — | — | — | 0.110 | 0.100 |
| Molybdenum | 2.56 | 2.20 | 0.63 | — | — | 0.58 | — | — | 0.118 | 0.108 |
| Ni-Tensyl | 2.80 | 2.51 | 0.68 | 1.71 | 0.54 | — | — | — | 0.101 | 0.092 |
| Manganese-nickel | 3.10 | 2.51 | 3.11 | 1.00 | — | — | — | — | 0.106 | 0.097 |
| Nickel-chromium | 3.41 | 1.03 | 0.65 | 1.49 | 0.54 | — | — | — | 0.116 | 0.106 |
| Ni-Resist | 2.41 | 1.80 | 0.62 | 13.70 | 3.37 | — | 6.41 | — | 0.081 | 0.075 |
| Aluminium-chromium | 2.70 | 0.96 | 0.58 | — | 0.95 | — | — | 7.00 | 0.079 | 0.072 |
| Nicrosilal | 1.81 | 6.42 | — | 18.65 | 2.02 | — | — | — | 0.070 | 0.063 |

combination with silicon have on the heat-conducting property of these materials. The general influence of nickel and silicon is to lower the conductivity, but where nickel and chromium or nickel and manganese are alloyed together, the influence of nickel in lowering the conductivity is counteracted to a slight degree by the chromium or manganese.

The influence of special elements in lowering the thermal conductivity of cast iron is seen to a more marked degree in the three heat-resisting irons. The Nicrosilal, with a high nickel and silicon content, has a value of 0.070 at 100° C., a value which is raised by 15% by replacing part of the nickel by copper and reducing the silicon. A somewhat similar value is obtained in the aluminium-chromium iron containing 7% of aluminium, and appears to indicate that aluminium, even in the presence of chromium, has a very pronounced influence in lowering the thermal conductivity of cast iron.

When the influence of special elements on grey cast iron is considered, it would appear that while silicon, nickel, aluminium, and copper lower the thermal conductivity due to the graphitising effect of these elements on cast iron, chromium, tungsten, and molybdenum raise the conductivity in proportion to the stability of their carbides. It has been suggested that the raising of the conductivity is produced by the carbide-forming elements causing the withdrawal of carbon from solution in the ferrite lamellae of the pearlite, the ferrite being the chief conducting phase, and the phase which must be looked to to account for variations in conductivity. It has also been suggested that the greater influence of nickel over silicon in lowering the conductivity, in spite of the greater graphitising power of silicon, may be due to the greater effect of silicon in precipitating carbon from solution in ferrite, thereby raising the conductivity in opposition to the direct influence of silicon in decreasing the conductivity as it enters into solution.

Thermal Conductivity of Some Ferrous and Non-Ferrous Metals

The influence of copper up to 3% on the thermal and electrical conductivities of grey cast iron has also been investigated by Lorig and Schnee⁷ on a series of six cast irons. Thermal conductivity determinations were made by heating one end of the test specimens and measuring the temperature gradient along the specimen and the quantity of heat passing through it by means of a heat-flow meter, which is a standard metal of known conductivity in contact with the cold end of the specimen. Values were obtained at approximately 160°, 240°, 315°, and 400° C. Electrical resistivities of the irons were measured by the potentiometric method over a length of 40 cm. at a temperature of 23.5° C. In Table III are given the values for the thermal conductivities of the various irons at 20°, 100°, 200°, 300° and 400° C., obtained by interpolation from the experimental data, together with the electrical resistivity values and composition of the various irons. The results obtained show the thermal conductivity to decrease up to 0.5% of copper then to return to normal with 2% copper after which it remains

TABLE III.

Thermal Conductivities and Electrical Resistivities of Copper Cast Irons (Lorig and Schnee).

| No. | Composition. | | | Thermal Conductivity — K. | | | | | Thermal Resistivity. |
|-----|--------------|------|------|---------------------------|-------|-------|-------|-------|----------------------|
| | C. | Si. | Cu. | 23° | 100° | 200° | 300° | 400° | |
| 30 | 3.16 | 1.54 | 0.00 | 0.112 | 0.110 | 0.106 | 0.103 | 0.099 | 58.4 |
| 31 | 3.29 | 1.50 | 0.53 | 0.105 | 0.103 | 0.100 | 0.098 | 0.095 | 61.4 |
| 32 | 3.18 | 1.59 | 0.99 | 0.108 | 0.106 | 0.103 | 0.100 | 0.098 | 61.8 |
| 33 | 3.15 | 1.58 | 1.45 | 0.108 | 0.106 | 0.103 | 0.099 | 0.096 | 61.0 |
| 34 | 3.18 | 1.49 | 1.98 | 0.114 | 0.110 | 0.104 | 0.099 | 0.093 | 59.2 |
| 35 | 3.16 | 1.44 | 3.10 | 0.112 | 0.110 | 0.105 | 0.102 | 0.098 | 57.4 |

⁷ C. E. Lorig and V. H. Schnee. Amer. Found. Assoc., 1940, Preprint No. 40-15.

constant, while the thermal resistivity increases slightly with copper up to 1% then decreases with further increases of copper, the resistivity of an iron containing 2.5% copper being similar to that obtained in a copper-free iron. The results obtained on the thermal conductivity are comparable with those obtained by the author.

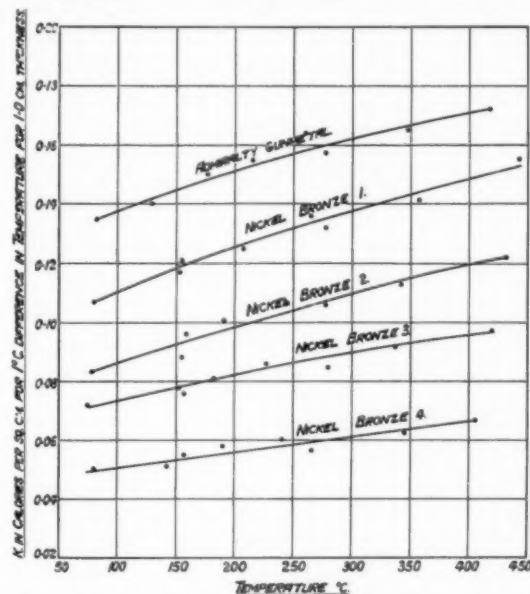


Fig. 2.—Thermal conductivities of nickel bronzes (Donaldson)

Copper and Nickel Alloys

To obtain reliable data on the thermal conductivity of some industrial alloys of copper and of nickel used for high-duty and other purposes, an investigation was carried out by the author* on some such alloys prepared according to definite commercial specifications. Thermal conductivities were determined over a range of temperature from 40° to 530° C., and on some of the alloys determinations were made in the cast and in the worked conditions of the material in order to determine the influence of work on the heat-conducting property.

The alloys tested were divided into three groups. The first group consisted of two high-tensile brasses of approximately the same composition, one cast and one rolled and annealed; two aluminium bronzes containing approximately 8.5 to 10.0% aluminium, 2% manganese, and the remainder copper, one cast and one rolled annealed; and a third cast aluminium bronze having the approximate composition aluminium 9.5%, manganese 2%, iron 3%, nickel 4%, and copper the remainder. The six alloys in the second group* were a cast Monel metal; a cold-rolled Monel metal; a K Monel metal, rolled and annealed, containing 2.50% of aluminium; two cast silicon-Monel metals containing 2.09 and 3.37% of silicon respectively; and a rolled Inconel, having an approximate composition of 80% nickel, 13% chromium, 6.5% iron. Four cast nickel bronzes were included in the third group, the first containing approximately 5% nickel, 5% tin, 2% zinc, copper the remainder; and the other three 10% tin, 1% zinc, with 15, 30 and 50% of nickel respectively, copper the remainder. Curves showing the thermal conductivities of these various alloys together with that of an Admiralty gunmetal (88 : 10 : 2) are given in Figs. 1 and 2.

All the alloys have a positive temperature coefficient, the conductivity increasing as the temperature increases, and the increases are regular and uniform. Between 100° C. and 400° C. the increase in conductivity varies from 22% in the rolled Inconel to 50% in the cast aluminium bronzes. The high tensile brass, aluminium bronze and Monel metal

*The names of the alloys in this group are registered trade-marks of the Mond Nickel Company, Ltd., London.
⁸ J. W. Donaldson. Journ. Inst. Metals, 1939 68, pp. 139-145.

have heat-conducting values which vary little in the cast and rolled condition of these materials, although the value for the rolled material is slightly greater in each case. It has been suggested that small variations in composition might account for the slight differences in conductivity, but differences for such variations do not seem probable, particularly in the case of the high-tensile brass and Monel metal, where differences in composition in the two conditions of the alloys are slight. Increase in conductivity in the worked condition is more probably due to denser and more compact structures.

The thermal conductivity of aluminium bronze containing approximately 90% copper, 10% aluminium, and a small proportion of manganese is about 20% less than that of high tensile brass, with 60% copper, 40% zinc, and somewhat similar to Admiralty gunmetal containing 88% copper, 10% tin, 2% zinc. The increase in conductivity with increase in temperature is more marked, however, in the aluminium bronze than in the gunmetal. The value of 0.138 obtained for the gunmetal at 100° C. is decreased to 0.111 at the same temperature by replacing 5% of the tin by 5% of nickel. The addition of 3% of iron and 4% of nickel to aluminium bronzes produces a decrease in thermal conductivity of about 30%, a value of 0.082 being obtained at 100° C., a value somewhat similar to that obtained in a nickel bronze containing approximately 10% tin, 15% nickel, 1% zinc, copper the remainder.

If the nickel in nickel bronze containing 10% tin, 1% zinc is increased from 15 to 30%, an alloy with a thermal conductivity of 0.074 at 100° C. is obtained. This value is about 15% higher than that obtained for Monel metal at the same temperature—namely, 0.065. The addition of 2.5% of aluminium to Monel metal reduces this value to 0.046, and a similar value is obtained by the addition of 3.4% of silicon. These latter alloys have also thermal conductivities approximating to that of a nickel bronze containing 50% of nickel. Inconel has very low thermal conductivity values of 0.037 and 0.045, being obtained at 100° and 400° C. respectively. It is of interest to compare these latter values with thermal conductivities of 0.055 at 600° C., 0.059 at 700° C. and 0.062 at 800° which have been calculated from electrical resistivity determinations on Inconel.

Thermal and Electrical Conductivities

In Section IX of the Second Report of the Alloy Steels Research Committee of the Iron and Steel Institute there are two papers dealing with the thermal and electrical

conductivities of irons and steels. The first by Powell and Hickman⁹ discusses the results obtained on a series of seven carbon steels seven alloy steels, and six high-alloy steels. The method used for the thermal conductivity measurements was that of longitudinal heat flow in a specially designed apparatus, and the electrical resistivity was determined from potential drop in the same specimen at the same time. Determinations for both properties are given at intervals of 50° C. from 0° C. to 350°, the values for 0° C. being extrapolated, as the lowest temperature at which thermal conductivity determinations were carried out was 40° C. and that of electrical resistivity determinations 18° C. In Table IV are given a brief summary of the compositions of the steels with their thermal conductivity and electrical resistivities at 50° and 300° C., together with the Lorenz function at those temperatures. The carbon and low-alloy steels were tested in their annealed condition, and the high-alloy steels were tested after suitable heat-treatments.

The thermal conductivities of the carbon steels at 50° range from 0.150 to 0.1075, and decrease as the carbon content increases. All the steels show a decrease in conductivity with increase in temperature, this negative temperature coefficient being most marked in the steels of lower carbon content, those of higher carbon content showing little or no change over the range 0° to 150° C. The values obtained for the carbon steels are considerably lower than those obtained by Esser, Eilender and Pütz for carbon steels (Table I), and are also slightly lower than the values obtained by the author,⁴ who found that the thermal conductivities at 100° C. of a series of carbon steel containing 0.10 to 1.09% carbon ranged from 0.161 to 0.118.

The thermal conductivities of the alloy steels range from 0.110 to 0.064 at 50° C., and those of the high-alloy steels from 0.065 to 0.033 at the same temperature. For the alloy steels of higher conductivity the temperature coefficients are negative over the range 50° to 300° C., but those of lower conductivity show positive coefficients. All the high alloy steels have positive temperature coefficients showing an increase in their heat-conductivity properties with increase in temperature. The results show in general that the more highly alloyed a steel is, the lower is its thermal conductivity, and that the decrease in thermal conductivity is accompanied by a transition from a negative to a positive temperature coefficient. The values obtained for the various steels and for their behaviour with temperature are also in general agreement with the values obtained

⁹ R. W. Powell and M. J. Hickman. Second Report Alloy Research Committee, Iron and Steel Inst.—Section IX, 1939, 242-251.

TABLE IV. THERMAL CONDUCTIVITIES, ELECTRICAL RESISTIVITIES, AND LORENZ FUNCTION OF CARBON AND ALLOY STEELS (POWELL AND HICKMAN).

| No. | Composition. | | | | | | | Thermal Conductivity. | | Electrical Resistivity. | | Lorenz Function. | |
|-----|--------------|------|-------|-------|-------|-------|-------|-----------------------|---------|-------------------------|---------|------------------|---------|
| | C. | Si. | Mn. | Ni. | Cr. | W. | Mo. | 50° C. | 300° C. | 50° C. | 300° C. | 50° C. | 300° C. |
| 1 | 0.06 | 0.01 | 0.38 | — | — | — | — | 0.150 | — | 14.6 | — | 0.68 | — |
| 2 | 0.08 | 0.08 | 0.31 | — | — | — | — | 0.140 | 0.118 | 15.9 | 34.7 | 0.69 | 0.715 |
| 3 | 0.23 | 0.11 | 0.635 | — | — | — | — | 0.1235 | 0.106 | 18.5 | 37.6 | 0.71 | 0.695 |
| 5 | 0.415 | 0.11 | 0.643 | — | — | — | — | 0.1235 | 0.109 | 19.3 | 38.4 | 0.74 | 0.73 |
| 6 | 0.435 | 0.20 | 0.69 | — | — | — | — | 0.115 | 0.105 | 21.2 | 40.0 | 0.755 | 0.735 |
| 7 | 0.80 | 0.13 | 0.32 | — | — | — | — | 0.118 | 0.099 | 19.8 | 39.7 | 0.725 | 0.685 |
| 8 | 1.22 | 0.16 | 0.35 | — | — | — | — | 0.1075 | — | 21.6 | — | 0.72 | — |
| 4 | 0.23 | 0.12 | 1.51 | 0.04 | 0.06 | Nil | 0.025 | 0.110 | — | 22.8 | — | 0.78 | — |
| 9 | 0.325 | 0.18 | 0.55 | 3.47 | 0.17 | Nil | 0.04 | 0.089 | 0.094 | 28.4 | 46.8 | 0.785 | 0.765 |
| 10 | 0.33 | 0.17 | 0.53 | 3.38 | 0.80 | Nil | 0.07 | 0.084 | — | 29.9 | — | 0.78 | — |
| 11 | 0.325 | 0.25 | 0.55 | 3.41 | 0.71 | Nil | 0.06 | 0.083 | 0.087 | 31.4 | 49.9 | 0.81 | 0.76 |
| 12 | 0.34 | 0.27 | 0.55 | 3.53 | 0.78 | Nil | 0.39 | 0.080 | 0.0855 | 32.4 | 50.6 | 0.805 | 0.755 |
| 19 | 0.315 | 0.20 | 0.69 | 0.673 | 1.09 | Nil | 0.012 | 0.114 | 0.101 | 22.9 | 42.0 | 0.81 | 0.74 |
| 20 | 0.35 | 0.21 | 0.59 | 0.26 | 0.88 | Nil | 0.20 | 0.102 | 0.097 | 24.3 | 43.3 | 0.78 | 0.735 |
| 21 | 0.485 | 1.98 | 0.90 | 0.156 | 0.04 | Nil | Nil | 0.064 | 0.0745 | 44.2 | 60.1 | 0.88 | 0.78 |
| 13 | 1.22 | 0.22 | 13.0 | 0.07 | 0.03 | Nil | Nil | 0.0335 | 0.043 | 71.5 | 92.4 | 0.74 | 0.695 |
| 14 | 0.28 | 0.15 | 0.89 | 28.37 | Trace | Nil | Nil | 0.033 | 0.042 | 86.8 | 102.0 | 0.89 | 0.75 |
| 15 | 0.08 | 0.68 | 0.37 | 8.14 | 19.11 | 0.60 | Nil | 0.0385 | 0.043 | 74.0 | 92.8 | 0.88 | 0.705 |
| 16 | 0.13 | 0.17 | 0.25 | 0.14 | 12.95 | Nil | Nil | 0.065 | 0.067 | 54.4 | 77.1 | 1.09 | 0.90 |
| 17 | 0.27 | 0.18 | 0.28 | 0.20 | 13.69 | 0.25 | 0.01 | 0.062 | 0.066 | 55.0 | 78.0 | 1.05 | 0.90 |
| 18 | 0.715 | 0.30 | 0.25 | 0.067 | 4.26 | 18.45 | Trace | 0.060 | 0.067 | 45.6 | 64.1 | 0.85 | 0.75 |

for alloy steels by Shelton¹⁰ and for tool steels by Hattori.¹¹

The electrical resistivities of the various steels are lowest for the carbon steels and highest for the high-alloy steels, and the temperature coefficient of resistance of each steel is positive. It follows, therefore, that the electrical conductivity of steel decreases as steel becomes more highly alloy, but differs from the thermal conductivity in that the temperature coefficient remains negative. With regard to the Lorenz function, which is the product of the thermal conductivity and the electrical conductivity divided by the absolute temperature and which, according to theory, should have the constant value of 0.585×10^{-8} if electrons are entirely responsible for both types of conduction, there are considerable departures from this value. Higher values are invariably obtained, and in most cases the departure is greatest at the lower temperatures and decreases with increase in temperature.

This relationship between thermal and electrical conductivity is also dealt with by Powell¹² in the second paper on conductivity in the Alloy Research Report. Existing data for the thermal and electrical conductivities of irons and steels are surveyed, and it is shown that, whereas at normal temperatures there is about a tenfold difference in their conductivities, at 800° C. most alloys of this type possess thermal conductivities lying between 0.05 and 0.08 cal./cm. sec. deg., and electrical conductivities lying between 8.0×10^3 and 9.5×10^3 reciprocal ohms per cm.³

10 S. M. Shelton, *Bureau of Standards, Journal of Research*, 1934, 12, 441-450.

11 D. Hattori, *Jour. Iron and Steel Inst.*, 1934, 129, 289-309.

12 R. W. Powell, *Second Report Alloy Research Committee, Iron and Steel Inst.*—Section IX, 1939, 253-268.

The Crystal Structure of Cementite, Fe_3O

THE structure of cementite has been the subject of two main investigations in the past. Westgren and Phragmén determined the unit cell, and from their data Hendricks found the space group and the positions of the iron atoms. With regard to the carbon atoms he did no more than suggest possible sites, without attempting to justify them on the basis of X-ray intensities. Westgren, however, pointed out that the carbon atoms on these sites would have to possess rather smaller radii than usual, and showed that there were alternative positions which would allow the introduction of carbon atoms of the normal size. He also showed that the X-ray intensities agreed rather better with his structure than with Hendricks', but was careful to state that the agreement was not sufficient to prove the correctness of the structure.

Both Hendricks and Westgren used only visual estimates of intensities, in which the successive classes differed by a factor of the order of 2. Since the carbon contributes only about 7% of the total number of electrons, the possibility that it may produce a difference of even one class in intensity is very remote. It is, therefore, essential that accurate measurements of intensity be used if the carbon atoms are to be located. Even with accurate intensities it was found that the carbon contributions would still be of the same order as the experimental error, and, moreover, of the same order as the changes due to the possible errors in the parameters quoted by Westgren. Thus, a completely objective survey of all the intensities is necessary in order to establish the carbon positions.

The best method of providing this survey is that of Fourier synthesis, which gives the electron density at every point in the unit cell directly from intensity measurements. From these measurements can be derived the numerical values of the structure amplitudes, but the signs of these structures amplitudes can be obtained only from a knowledge of the positions of the iron atoms. No

At low temperatures many of these irons and steels deviate considerably from the Wiedemann-Franz-Lorenz Law, but with increase in temperature these deviations diminish, and between temperatures of 500° and 800° C. the values so far determined for the Lorenz function, when the conductivities are expressed in the above-mentioned units, all lie between 0.64×10^{-8} and 0.77×10^{-8} . When the thermal conductivity K is plotted against the product of the electrical conductivity σ and the absolute temperature T , the majority of the points fall within 12% of the line represented by $K = 0.625 \times 10^{-8} \sigma T + 0.006$. Considerable departures occur, however, for cast iron containing graphite and for one variety of stainless steel, and it is probable that further experiments on other irons and steels may reveal other exceptions. In general, however, the curve furnishes a means of predicting to within about 10% the value of the thermal conductivity at any temperature from a measurement of the electrical conductivity at that temperature.

Using this curve, Powell, having determined the electrical resistivities of ingot iron, when heated in an evacuated enclosure to 1,400° C., gives values for the thermal conductivity of such iron at 800°, 1,000°, 1,200°, and 1,400° C. as 0.079, 0.075, 0.082, and 0.088. Subsequent determinations by the same investigator have shown that the thermal conductivity of iron falls to 0.065 at 900° C. and rises to 0.0665 at 1,000° C. The Lorenz function is found to decrease rapidly beyond the magnetic transformation point, and at 1,000° C. has a value of 0.60×10^{-8} , and this decrease leads to thermal conductivity values for gamma iron which are some 12% less than those given above.

assumptions, however, need be made about the positions of the carbon atoms, since any structure amplitude which could be reversed in sign by the carbon contribution would give an intensity too small to be observed. Thus, the results obtained must be quite independent of any preconceived ideas about the positions of the carbon atoms.

Although Fourier methods can give the electron density in three dimensions, it has been usual to reduce the formulae to two dimensions in order to simplify the calculations. This then gives the projection of the electron density on a plane. In the case of cementite this would mean that a projection would contain 12 iron atoms and 4 carbon atoms, and thus the latter might be very difficult to detect. If, however one sums the three-dimensional series, one can be sure that none of the atoms will come so close to another that it cannot be observed.

As is shown in the present paper by Dr. H. Lipson and N. J. Petch, which will be presented for discussion at the September meeting of the Iron and Steel Institute, it is not necessary to sum the series at every point in the unit cell in order to establish the positions of the carbon atoms, so that the work can be kept within reasonable bounds.

Three-dimensional Fourier synthesis requires the observation of all the intensities within a given value of the Bragg angle, θ . The most convenient way of observing these is by the Debye-Scherrer powder method. The main objection to this method is the lack of resolution of the spectra, but since the unit cell of cementite is not large this objection is not serious, and in the work described in the present paper it was overcome to a considerable extent by using a large camera, of diameter 35 cm.

By direct calculation of the electron density in two different planes in the unit cell, using accurately measured intensities of X-ray reflection, as indicated above, the present authors have found the position of the carbon atoms. They agree very well with the more probable of the two sets of positions previously suggested. The authors also give a description of the structure, with a discussion of its possible relation to the structures of ferritic and austenitic iron.

Some Recent Results when Investigating the Roasting of Magnetised Brown Iron Ore

AN experimental plant for magnetising iron ore has been built in the Institute for Iron and Steel Research, with the object of making detailed investigations of a new method, developed in the Institute, for enrichment of poor red and brown iron ores before further treatment. The plant, which is described by Luyken and Kremer*, consists of two kilns, one for roasting and the other for cooling, through which the material travels. Both are rotated by means of a motor, and the heating of the upper or roasting kiln is effected by low-pressure gas. In this plant three ores were investigated, each containing about 24% iron, but otherwise of different analysis. The results are given in detail in several tables which indicate that the method can be considered to be satisfactory; for example, in one case, with ore originally containing only 17% iron, a concentrate could be obtained containing 49.6% iron, from which an iron-yield up to 86% could be obtained.

In order to determine whether magnetising roasting can be considered as superior to other methods, the investigated ores were dressed by these other methods, and it could be concluded that the roasting gives a considerable improved enrichment. In addition, the method showed other advantages—i.e., it is independent to a great extent of the physical and chemical conditions of the ores and of differences in the roasting and cooling temperatures; especially, however, the decrease of contents of arsenic should be mentioned. This was found for two different ores, by roasting as well as by the magnetic separation methods—i.e., 18% and 28% by roasting and about 24% by magnetic separation. It is pointed out that, as a result of the losses of iron during the magnetic separation, the total decrease of arsenic is not important; but, on the other hand, it should not be forgotten that this feature of the method is effected without increase of costs.

The Influence of Chemically-Combined Water on the Fuel Consumption when Sintering Iron Ores

Further investigations have been carried out to determine whether brown iron ores or their concentrates could be sintered more economically than roasted concentrates. These investigations, which are described by the same authors*, were carried out in a small laboratory sintering plant. It was found that when sintering a roasted concentrate, a more solid sinter, as well as a smaller quantity of unsintered material, was obtained with equal fuel consumption.

This result should be checked in actual production on a Dwight-Lloyd plant. Two different ores could be investigated, a brown and a red iron ore. The first one needed a smaller addition of fuel, but this difference was caused by the variety of the two ores, and it was clear that the influence of the chemically-combined water on the fuel consumption could be determined only if the same ore was investigated in the roasted as well as in the unroasted state.

In actual production sufficient quantities of ores for such investigations could not be obtained, and it was necessary to revert to experiments in the Iron Research Institute, where sufficient quantities of concentrates could be produced in a rotary furnace. The results of these investigations showed that the sinter of brown iron ores was less stable than that from a roasted concentrate. Moreover, considerably less fuel was consumed for the latter, and a smaller proportion of unsintered material obtained.

It can therefore be concluded from these investigations that the extraction of chemically-combined water from brown iron ore during sintering increases the consumption of fuel.

* Mitteilungen aus dem Kaiser-Wilhelm Institut für Eisenforschung, Vol. XX, No. 21.

Behaviour of Steels Under High Temperature when Exposed to Alternating and Comprehensive Stresses

PUBLISHED work on the behaviour of steels at high temperature when undergoing alternating stresses is meagre and incomplete; investigations have been concerned with repeated stresses—i.e., from zero to a maximum of tension. A recent investigation, however, reported by Hempel and Ardelt* deals with alternating tensile and compressive stresses.

The heating of the test-pieces was by electric air furnace, and a hydraulic testing machine was used capable of applying a load up to 60 tons for static tensile and compressive tests, from zero to +40 tons, respectively, for repeating tensile, compressive and bending tests, and up to +20 tons for alternating tensile, compressive or bending tests. Preliminary tests were necessary to solve problems of procedure, and in regulating and distributing the temperature over the length of the test-pieces. A new device was developed by which the rate of elongation could be determined and automatically recorded during a period of 10 hours with an accuracy of $\pm 0.5 \times 10^{-4}\%$ per hour. Three types of steel were treated—(1) an annealed steel containing 0.58% C.; (2) a carbon-molybdenum steel containing 0.14% C. and 0.51% Mo; and (3) a steel containing 0.56% C, 15.5% Cr, 3.3% Ni and 2.02% W.

The course of elongation at 500° C. was observed at static tensile and compressive stresses, as well as under alternating tensile-compressive stresses for various average tensile strengths. The time-elongation curves were of similar shape in all cases; they showed a rapid elongation during the first hours of the experiments, and afterwards a decrease in the rate of elongation.

In all cases the relation of the total and of the permanent elongation, as well as of the rate of elongation to the various mean tensile strengths, were determined at constant maximum strengths. It is remarkable that even for the mean strength zero—i.e., for equal positive and negative maximum strengths—an elongation takes place. During static tensile and compressive tests it was found that the time-elongation curves of the tensile tests on the carbon-molybdenum steel were considerably higher than those of the compressive tests, equal load being provided in both cases.

Finally, the probable course of the fatigue strength curves at 500° C. was determined. As delimitation of the diagrams not only were the 0.2 limit and the creep limit used, but also the permanent elongations under alternating stress and the rates of elongation.

* Mitteilungen aus dem Kaiser-Wilhelm Institut für Eisenforschung Zu Düsseldorf, Vol. XXI, Nr. 7.

The Nickel Bulletin—July, 1940

AN historical survey of the electrolytic pickling and polishing of metals, which is abstracted in the current issue of *The Nickel Bulletin*, shows that this subject has received much attention during recent years. Bright plating, another subject of wide interest, is leaving the experimental stage; a large scale bright-plating plant is now in operation in California. Some phases of the various processes are still being investigated—for instance, extensive work on the value of organic compounds as brighteners has been carried out on the Continent.

Considerable attention is given to a paper which deals with the properties and uses of hardenable nickel-bronze casting alloy, while shorter abstracts deal with the constitution of iron-rich nickel-iron alloys, cobalt and copper additions to nickel-aluminium-iron magnet alloys, alloy cast iron in railway equipment, improvements in hand tools, etc. A few interesting hints are given on the cleaning of stainless steel, details are also given of Monel nails, which are claimed to be superior to screws in holding power.

The Iron-Plating of Specimens for Microscopical Examination*

By E. A. Jenkinson, B.Sc.

An account is given of a method of supporting fractured surfaces of creep and tensile test pieces, so that longitudinal sections can be prepared for microscopical examination right up to the fracture. After a brief discussion of the disadvantages of a fusible alloy mounting and of a thick copper electro-deposit, experimental work on the production of heavy electro-deposits of iron with suitable qualities is described. The general effect of temperature and current density on the microstructure of the plating is illustrated by micrographs. It has been found possible to obtain suitable deposits when employing a temperature of 85° C. and current densities between 5 and 20 amp. per sq. ft.

IN the investigation of the deformation of metals at high temperatures¹ longitudinal sections of fractured test-pieces were examined microscopically to determine the type of fracture and the changes in structure close to the fracture and along the length of the test-piece. It was found necessary to give some form of support to the fractured surface so that flat polished sections could be prepared and examined right up to the fracture.

A fusible-alloy mounting was unsuitable for two reasons. Its softness in comparison with the test-piece made it difficult to retain a truly flat surface in polishing, and the specimen in the vicinity of the fusible metal was not etched so easily as the rest of the material. A thick electro-deposit of copper was next used. Some previous workers^{2, 3, 4} in the Metallurgy Department, National Physical Laboratory, had employed electrodeposited copper or a combination of electrodeposited copper and iron for the protection of surfaces and fractures. Copper was readily deposited and heavy layers were easily obtained on the specimens. Experience with this method led to the conclusion that it possessed disadvantages similar to those already found when using fusible metal, but to a somewhat less extent. There were also the additional disadvantages that the adherence of the copper plate was not always good, and that two solutions were needed: First an alkaline cyanide bath at a low current density and then an acid sulphate one, which could not be used directly on the iron specimen owing to chemical action.

Iron deposition was then tried, and after some experimental work it was found possible to plate on to fractures a heavy electro-deposit of iron which had suitable qualities for the subsequent microscopical examination. The qualities required were: (a) Toughness, (b) firm adherence, (c) absence of gas cavities, (d) uniform thickness of approximately 0.020 in., and (e) ability to permit uniform etching of the specimen. The present paper gives an account of

the experiments carried out to obtain this type of deposit.

Experimental Work

The electrolytic bath used was the same as that employed by Cain, Schramm and Cleaves⁵ and Tritton and Hanson⁶ in the preparation of electrolytic iron, and consisted of

| | | |
|------------------|---------|------------|
| Ferrous chloride | | 288 g. |
| Sodium chloride | | 57 g. |
| Distilled water | | 1,000 c.c. |

A special grade of ferrous chloride containing 0.0032% of nickel, 0.0076% of manganese and 0.001% of phosphorus was employed together with sodium chloride of "Analar" purity. Solutions were prepared by adding the ferrous chloride and the sodium chloride to distilled water, and allowing them to dissolve cold. Small amounts of ferric chloride were removed by filtering before use. At the high temperatures at which the bath was operated loss by evaporation was a serious factor, and a device was used to maintain the bath at constant level by the addition of distilled water.

The plating equipment consisted of a glass vessel of 400 c.c. capacity provided with a small electric hot-plate to maintain the temperature. The specimen formed the cathode, which was suspended from a spindle rotating at approximately 50 r.p.m. The anode consisted of a piece of Armeo ingot iron plate, 4 × 1 × ½ in., bound with linen tape to prevent sludge from entering the bath. The following ranges of conditions were examined:

1. Current density, 5 to 35 amp. per sq. ft. of cathode.
2. Temperature, 70 to 100° C.

Specimens plated were fractured halves of creep or tensile standard round test-pieces. In most cases the tests had been conducted *in vacuo*, the fractures being bright, but in a few cases tests had been made in air, and the fractures were then more or less badly oxidised. Materials plated included Armeo ingot iron, high-purity iron and steels containing from 0.4% to 1.1% of carbon. Specimens

* Communication from the National Physical Laboratory to the Iron and Steel Institute for discussion at the September meeting, 1940. (Advance copy.)

¹ Jenkins and Mellor: *Journal of the Iron and Steel Institute*, 1935, No. II., p. 179.

² Rosenhain: *Proceedings of the Royal Society*, 1905, vol. 74, p. 557.

³ Rosenhain: *Journal of the Iron and Steel Institute*, 1906, No. II., p. 196.

⁴ Rosenhain and Humphrey: *Journal of the Iron and Steel Institute*, 1913, No. I., p. 219.

⁵ Cain, Schramm and Cleaves: *Bulletin of the Bureau of Standards*, 1916-17 vol. 13, p. I.

⁶ Tritton and Hanson: *Journal of the Iron and Steel Institute*, 1924, No. II., p. 90.

Fig. 1.—1.1% carbon steel, plated at 70° C., c.d. 12 amp. per sq. ft. × 150.

Fig. 2.—1.1% carbon steel, plated at 75° C., c.d. 26 amp. per sq. ft. × 500.

Fig. 3.—Armeo Ingot iron, plated at 75° C., c.d. 7.6 amp. per sq. ft. × 150.

Fig. 4.—High-purity iron, plated at 75° C., c.d. 15 amp. per sq. ft. × 150.

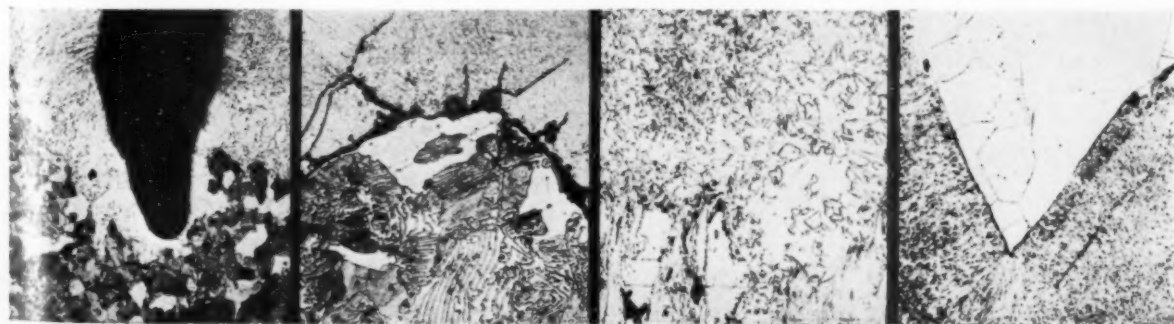




Fig. 5.—1.1% carbon steel, plated at 80° C., c.d. 9 amp. per sq. ft. $\times 500$.

Fig. 6.—0.57% carbon steel, plated at 80° C., c.d. 13 amp. per sq. ft. $\times 500$.

Fig. 7.—1.1% carbon steel, plated at 85° C., c.d. 6 amp. per sq. ft. $\times 500$.

Fig. 8.—0.86% carbon steel, plated at 85° C., c.d. 19 amp. per sq. ft. $\times 500$.

were plated in the condition in which they were removed from the testing unit without washing, cleaning or handling in any way.

The fractures of most of the test-pieces were very irregular, but in the case of a few of the high-temperature creep-test specimens (at 850° C. and 950° C.), the fractures were smooth and rounded. Examples of this type are given in the photomicrographs (Figs. 2, 8, 9 and 11).

Several hundred heavy iron deposits were prepared. The general effect of temperature and current density on the micro-structure of the plating is illustrated by Figs. 1 to 12. The electro-deposit forms the upper portion of the micrograph in every case.

Description of Microstructures

Iron deposited at temperatures below 70° C. took up large quantities of hydrogen, giving a very rough, brittle and friable electrodeposit. Some of the hydrogen released as plating proceeded remained in contact with the cathode, and ultimately cavities were produced where these gas bubbles remained (Fig. 1). Gas cavities could be produced at higher temperatures, but only at high current densities and even then not to the same extent (Fig. 10). Brittleness in a deposit led to internal cracking (Fig. 2). Between 70° and 80° C. the deposits obtained at current densities of 7.5 to 26 amp. per sq. ft. had a very fine grain size (Figs. 1 to 4), but suffered from excessive numbers of gas cavities, most of which, however, were small. At temperatures between 80° and 100° C. the deposits were fine-grained and free from cavities except at the highest current densities employed (Fig. 10). The line of demarcation between electrodeposit and specimen was very slight, and specimens in the unetched condition showed none: on etching, the different structures of the basis metal and the electrodeposit gave adequate definition between the two. An example of this is given in Fig. 5, where the basis metal consists of a spheroidised steel with a ferrite matrix.

In general, adherent deposits were obtained irrespective of surface conditions. Surface oxidation of the original

material tended to increase the risk of gaseous cavities at the higher temperatures, but, provided that low current densities were employed, smooth deposits free from gas cavities were obtained (Fig. 6). In several cases where heavy oxidation had taken place adherent smooth deposits were formed. A good deposit was obtained most readily with a rough bright surface (Figs. 3, 7 and 9).

Figs. 11 and 12 illustrate deposits produced at 95° C. with relatively high current densities. They were not so free from inclusions as those formed at 85° C.

Columnar crystals were almost entirely absent from the deposits in the present work, except perhaps in those formed at 95° C. (Fig. 12). Probably at the low current densities employed the grains were formed sufficiently slowly to permit more or less uniform growth.

Firmly adherent deposits were obtained at all current densities (5 to 35 amp. per sq. ft.) between temperatures of 70° and 100° C., except in the case of a specimen plated at 75° C. with a current density of 26 amp. per sq. ft. (Fig. 2). This was a special case, in that the fracture was blunt, and not rough and jagged. At higher temperatures and lower current densities, however, blunt fractures were successfully plated (Figs. 8 and 9).

Conclusions

It has been found possible to obtain uniform and adherent deposits free from gas cavities both on bright test-pieces and on those which have suffered some degree of oxidation.

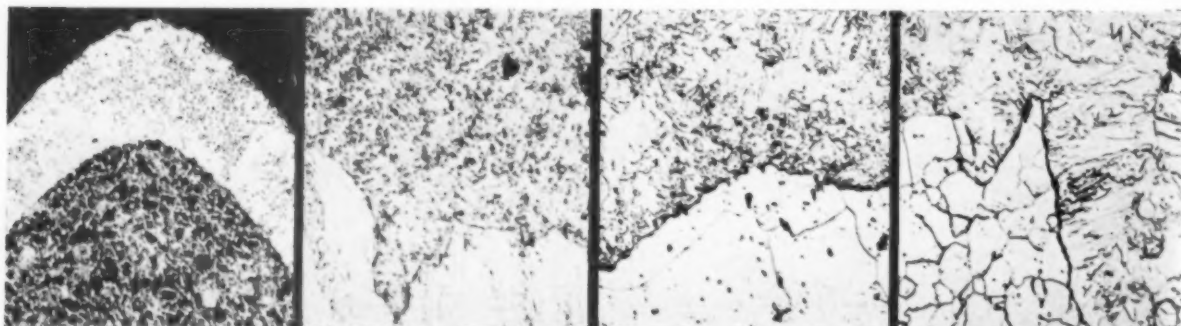
Plating at temperatures below 80° C. tends to produce gas cavities, brittleness and internal cracking of the deposit. At 85° C. the optimum conditions exist and the deposit is tough, very adherent and free from defects such as gas cavities. At higher temperatures the bath is more difficult to operate, owing to excessive evaporation, and the deposits are somewhat coarser and less clean than those produced at 85° C. It has been found that high current densities tend to produce gas cavities and fine grain size with consequent brittleness. The best conditions of current density are found to be from 5 to 20 amp. per sq. ft.

Fig. 9.—0.57% carbon steel, plated at 85° C., c.d. 9 amp. per sq. ft. $\times 50$.

Fig. 10.—High-purity iron, plated at 85° C., c.d. 27 amp. per sq. ft. $\times 500$.

Fig. 11.—High-purity iron, plated at 95° C., c.d. 22 amp. per sq. ft. $\times 150$.

Fig. 12.—High-purity iron, plated at 95° C., c.d. 34 amp. per sq. ft. $\times 500$.



Heat Treatment in Relation to Forming of Aluminium Alloy Sheet

The successful forming of high-strength aluminium-alloy sheet is largely dependent upon its condition at the time the operations are to be performed. Sheet is generally formed in the solution-treated condition, and in this article heat-treatment technique is discussed; the importance of temperature control is emphasised, and particular attention directed to the lapse of time between heat-treatment and the forming of alloy sheet which age-hardens at atmospheric temperature.

H EAT-TREATMENT techniques devised for high-strength aluminium alloys not only enable a desired combination of mechanical properties ultimately to be achieved, they also have a considerable bearing on forming procedure adopted with the materials, particularly in the form of sheet and strip.

The means whereby heat-treatable aluminium alloys may be improved in mechanical strength are now well known—i.e., a solution treatment at a temperature in the neighbourhood of 500° C., as a result of which the material is left in a comparatively soft condition, followed by subsequent hardening which may occur as a result of "spontaneous" (natural) ageing, or, as the result of an additional low-temperature precipitation treatment. Details of chemical condition, heat-treatment temperatures and mechanical properties for some high-strength sheet alloys are given in Table I.

Salt baths, generally utilising a eutectic mixture of sodium and potassium nitrates, are widely used for solution treatment. For the precipitation hardening treatment air furnaces are usually preferred, although for small work low-melting-point eutectic mixtures of salt or oil baths are sometimes used. For annealing, which involves a period of soaking at a temperature generally in the range 340° C. to 420° C., forced-air circulation furnaces—heated by electricity, gas, or oil—are the most commonly-used equipment.

After solution treatment, alloys of the "spontaneous ageing" group remain in the soft condition for a short period, after which hardening occurs at room temperature, and is completed in four to five days. Advantage may be taken of the temporary condition of softness for carrying out cold-forming operations which would be impossible with the material in the fully-aged condition. The period following the solution treatment in which such forming may be carried out is generally restricted to approximately one hour, although by resorting to refrigeration storage, to which reference will be made later, the period may be considerably extended.

In the use of alloys which require a further low-temperature treatment (double heat-treatment alloys) to develop maximum strength, spontaneous hardening does not occur to anything like the same extent. In fact, after a short period following the solution treatment, the material reaches a stable condition in which it is still possible to carry out relatively severe forming operations. The subsequent precipitation hardening treatment consists in heating the material at a temperature of about 170° C. for a period ranging between 10 to 18 hours.

The convenience which attaches to forming sheet in the solution-treated condition is well recognised in the aircraft industry. By avoiding a need for annealing, considerable savings of time are effected and economic planning of production rendered much simpler. Certain sheet materials belonging to the double heat-treatment class, which remain stable without age-hardening for an indefinite period, can be dispatched to the works in the solution-treated condition, this, of course, resulting in additional economy of time and equipment. When we consider that 64·2% by weight

of the airframe of an Empire Flying Boat (9,700 lb. out of a total weight of 15,100 lb.) consists of aluminium-alloy sheet, the importance of simplification in works procedure becomes very evident.

For very severe forming operations it is generally advisable to anneal, since it is in the annealed form that aluminium alloys possess their greatest ductility. It is, however, quite safe to carry out very appreciable amounts of forming in the solution-treated condition provided that too frequent reheat-treatment, particularly with aluminium-coated sheet, be avoided, and that essential precautions laid down by light-alloy sheet manufacturers with regard to treatment temperatures and temperature measurement, times of immersion and disposition of the load in the bath be fully observed.

The relatively narrow limits of heat-treatment temperature ranges for high-strength alloy systems, invariably within $\pm 10^\circ \text{C.}$ of a stipulated temperature, demand that the greatest attention be paid to temperature control of salt baths used for solution treatment. Pyrometric equipment is now available to meet these stringent requirements to a thoroughly satisfactory degree.

Pyrometers of the base-metal thermocouple type are generally preferred, a common feature being the use of thin refractory sleeves in place of the heavy metal sheaths formerly employed. This latter type of construction tended to render the instrument somewhat insensitive to temperature variation. In one test of a metal-sheath pyrometer against an accurate standard, it was found that the former was lagging behind the latter to the extent of 45° C.

The sheet undergoing reheat-treatment, in this case Alcad to DTD.275, a material widely used in aircraft construction, was, as a result of relying upon the inaccurate instrument, overheated and later showed signs of cracking. Substitution of the metal sheath by thin asbestos sleeves resulted in the avoidance of any repetition of the trouble. The example illustrates the desirability of carrying out a periodic check of pyrometers against accurate thermocouples of the bare-wire type. Dangers of over-heating, the most common cause of tendencies to crack during forming, are thereby avoided.

Errors resulting from pyrometer lag are aggravated if the salt be allowed to solidify overnight. This practice does not result in the economy it is sometimes supposed to achieve, for both time and heat necessary for remelting are considerable. The bath should be kept at a temperature just above the melting point of the salt, so that it may be raised rapidly to operating temperature when required. The provision of efficient insulation for the framework of the bath also assists considerably in the minimising of heating costs.

Automatic temperature-control and temperature-recording equipment are both desirable in solution-treatment plant. The former ensures against the temperature passing out of the stipulated range, whilst the latter provides continuous records of bath conditions. These records are, of course, of considerable value in investigations where it is necessary to trace the history of a batch of material

from the light-alloy factory to the stage where components produced from it are incorporated in aircraft or other structures.

The period for which sheet must be immersed in the salt naturally depends on the gauge of the material. A 22-gauge sheet requires a period of 10 to 14 min., but for three-gauge material a soaking period of 30 to 38 min. is necessary. Intermediate gauges require immersion for appropriate periods between these limits. Soaking for less than the required time will inevitably result in cracking after a limited amount of forming.

It is also important that the sheets be suspended independently in the bath to ensure uniformity of temperature throughout each sheet. Failure to prevent contact between sheets in the bath will result in ineffective heat-

until they are required, and the disadvantages attendant upon maintaining a supply of freshly-treated rivets avoided.

The method is now being extended to deal with sheet. Investigations have revealed that a storage temperature of -6°C . to -10°C . will prevent any appreciable age-hardening in the initial 50 hours of storage. Due to the bright surface of aluminium sheet, cooling of parts to the storage temperature takes some time, during which age-hardening might occur. Consequently, it is considered desirable before storage to quench the material in a liquid quench both situated inside the refrigerator. The development of refrigeration storage is of considerable interest, and its progress will undoubtedly be closely followed by all concerned with the forming of aluminium-alloy sheet.

TABLE I.

| Alloy. | Type. | Chemical Composition. | | | | | Solution Treatment. | Hardening. | Mechanical Properties. | | |
|----------------------------------|-------------------------|-----------------------|-----|-----|-----|-----|---|--|--|----------------------------|--------------------------|
| | | Si. | Cu. | Mg. | Mn. | Fe. | | | 0.1% Proof Stress. Tons per Sq. In. | Ultimate Tensile Strength. | Elongation % on 2 in. |
| NA 17.8 to BSS. 5L3 | Spontaneous Ageing.... | — | 4.0 | 0.6 | 0.6 | — | $490 \pm 10^{\circ}\text{C}$. water-quenched. | Five days' natural ageing. | 15 min. | 25 min. | 15 min. |
| NA. 21s to BSS. L47 (Alclad). | Double Heat-treatment.. | 0.8 | 4.0 | 0.4 | 0.8 | 0.7 | $510 \pm 10^{\circ}\text{C}$. Water-quenched. | Not less than 10 hours' heating at 165°C . | 19 min. | 25 min. | 8 min. |

Typical high-strength aluminium-sheet alloys.

Chemical compositions nominal.

Mechanical strength figures in heat-treated condition.

treatment at the points of contact and ultimately lead to localised cracking at these points during forming.

The requirement that, in the case of "spontaneous ageing" alloys, solution-treated parts must be formed within approximately one hour after treatment has undoubtedly led to trouble in the aircraft industry. Unless the most careful co-ordination exists between the solution-treatment plant and the presses a fall off in output is likely to occur.

If full advantage is not taken of the hour available, reheat-treatment may be necessary before even relatively small amounts of forming are completed. In the case of severe forming, the number of reheat-treatments necessary before completion may be raised to an excessive figure, this resulting in excessive grain growth with all its attendant problems.

As an alternative to reheat-treatment, works personnel may attempt to complete forming with the material in too hard a condition, thereby running the risk of cracking. In all cases, it will be observed, failure to carry out the maximum amount of forming within the hour results in increased costs and may, in the worst case, result in the material becoming scrap.

To attempt to avoid the problem by an unnecessary extension of annealing practice again raises costs, and, what is of extreme importance at the present time, retards output. The problem demands careful planning to ensure that freshly-treated blanks are passed immediately to the presses; such is the only way of obtaining maximum press efficiency and avoiding thereby a serious bottleneck in production.

Mention has been made of refrigeration storage. For some time now it has been the practice in the aircraft industry to store duralumin rivets at sub-normal temperatures, this to take advantage of the fact that at low temperatures the soft condition following upon solution treatment may be considerably extended. By this means the rivets may be retained in a condition fit for driving

Antifriction Alloys

THE effect of other metals on the working properties of aluminium bronzes used as antifriction alloys are discussed by Tschernajev and Makuschenko in *Foundry* (U.S.S.R.), vol. 10, No. 8, page 20-23. Over 5% iron increases the tensile strength and reduces elongation: working qualities are improved by one to 1.5% lead; a tin content of 0.1% is detrimental. Aluminium bronzes are considered to be at least the equal of tin bronzes as antifriction alloys.

Wild-Barfield's New Works

THE growing demand for electric heat-treatment furnaces and the difficulty of providing increased manufacturing facilities adjacent to the present works of Wild-Barfield Electric Furnaces Ltd., have caused this firm to obtain a new site in Hertfordshire, on which new works and offices are being erected. Work had started on this site last September, but circumstances prevented the erection of the building so quickly as was originally intended, steady progress, however, has been made and although the works may not be quite ready for occupation on the date specified, the offices are completed, and the staff of the designs and drawing offices, and accounts, sales and executive departments are in possession. In future the permanent address is Elecburn Works, Watford By-Pass, Watford, Herts. The telephone number is Watford 6094 (4 lines), and telegraphic address, "Elecburn, Watford."

The new buildings are designed on modern lines, of brick and reinforced concrete construction. Direct communication from offices to works is provided, and to the new buildings also completed which house the research department and laboratory with full chemical and physical testing equipment. The general layout has been designed to permit of extensions to give double the present area, spare ground on the main frontage being available for this purpose. Manufacturing and handling facilities in the bays at present under construction will enable production to be accelerated.

The Problem of Fatigue Fracture:

Magnetic Powder Pictures and Fatigue Fracture Cracks

An investigation is discussed on the formation and propagation of fatigue cracks in steel by subjecting specimens to alternating bend tests and examining them by the magnetic powder process. A comparison is made with the results obtained by other investigators.

FATIGUE fractures can take place on parts of machinery or structures when alternating stresses, after exceeding a certain load, coincide with times of stresses of some length. These fractures can be recognised by the fine grain and the arc limitation of the fracture surface, while the remaining section shows coarse grain and deformations on the edge of the test piece. Changes from one section to another or defects of various kinds as well as excessive changes of structure under alternating stress, generally increase the danger of fatigue fractures. Much work has been carried out with a view to increasing knowledge of fractures of this type, and many investigations are reviewed by Hempel¹, who also discusses the results of his work on the formation and propagation of fatigue cracks in steel by subjecting specimens to alternating bend tests which were examined by the magnetic powder process.

It is generally recognised that, at the beginning, the material is displaced at slip planes of a few crystals, but with the increasing duration of the test the number of these surfaces increase, and, when higher loads are used, even the displacement of crystals takes place, i.e., forces operating on some parts of the material disturb the lattice structure by local transformation and by external loads.^{1, 2} If these disturbances increase in strength as well as in extent the deformation capacity at these places will be exhausted by degrees, and finally a crack will appear. Such a crack has the same effect as a notch, and the result, consequently, is an increase of the stress at the bottom edge of the crack.^{3, 4} a further cracking of the material, and finally fracture of the remaining section.⁵

In order to determine the cause of fatigue fractures, tension-compression-fatigue tests have been carried out on nickel-steel specimens about their fatigue strength with and without intermediate annealing.⁶ It was found that the fatigue strength was lowered by the annealing, and that the fracture started when the reversals of the load were about one-sixth of the figure necessary for the fracture when the load remains equal; the intermediate annealing did not remove the damage that had taken place.

Bending tests on specimens with advance loads of alternating stress in fluid air showed the shearing strength.^{7, 8} According to these investigations, the cause of the fatigue fracture is disorganisation of the cohesive power. It is further known that the ratio between the surfaces of the vibratory fracture and of the force fracture changes according to material, extent and kind of the stress, as well as the form of the test-piece or machine part. In a series of experiments 11 different forms of test-pieces have been investigated, and as the fractures showed a more or

less good approach to geometrical forms, concentric, eccentric, double-convex and concave-convex fracture surfaces could be distinguished.⁹

The spreading of the fracture surface has been investigated in detail by heating the test-pieces after certain periods of alternating stress up to 250° to 400° C.; the fracture surface was thus divided into various zones of annealing colours. No doubt the progress of the fatigue fracture per unit of time could not be determined by this method, but the "rule of the concave spreading of the fatigue fracture" could be established for rotation bending tests. It shows that there is a certain ratio between the depth and the angle of the crack.¹⁰

Investigations on the rate of spread of a fatigue crack were based on the fact that the material had remained unstressed for some time after being subjected to alternating stress. A darker colour appears at the place where the cracked surface passes over into the sounder material. Plotting the depth of the crack as abscissae and the average rate, i.e., the ratio of the distance of two crack lines to the number of load alterations as ordinate, it has been found for alloy steels of 70 to 120 kg/mm.² (44.5 to 76.5 tons/eg. in.) tensile strength, and alternating loads of 35 to 50 kg/mm.² (22.5 to 32 tons/eg. in.), that the cracking rate is a square function of the progress of the fracture; the rate of the fracture progress could not be determined definitely by these investigations.

In some of these experiments it was concluded that the crack progresses with constant speed and the number of load alternations between the beginning of the crack, and the fracture was only 6.6% of the number of reversals after which the final fracture would take place. Other investigations seemed to show that the fatigue fracture begins about one-sixth of this number of reversals; but the kind of load was different; in the first case bending alternating stress was applied, and in the second tension-compression alternating stress connected with constant deformation. Whether these results are due to different kinds of stress, various loads about the fatigue strength, or to variation of material, cannot be stated, but one aspect should be taken into consideration: If a test-piece is under constant load, the actual working section decreases continuously after the cracking has started; consequently, the specific load of the remaining section increases constantly, and the progress of the crack is more and more accelerated. If, however, the deformation of the test-

piece is kept constant, the "spring-coefficient" $\frac{E}{C=c/q}$ decreases with increasing depth of the crack, E meaning modulus of elasticity, c the length and q the section of the test-piece; consequently, the load also decreases continuously—i.e., the cracking will proceed much slower. This has been confirmed by an investigation on flat test-pieces, which were one-sidedly clamped and exposed to alternating bending stress under constant deformation. The cracking proceeded here also about proportionally to the square of the depth of the crack, but there were great numerical differences from the results of the above-mentioned investigations.

Further investigations have been made on wagon-

¹ Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1929, **21**, 9, pp. 147-162.

² R. Mailänder: Ermüdungserscheinungen und Dauerversuche. Werkstoffausseh. 1921, **30**, p. 14. Stahl Eisen.

³ H. J. Gough: Crystalline Structure in Relation to Failure of Metals—especially by Fatigue. Proc. Amer. Soc. Test. Mater., **33**, 1933, II, pp. 3/114. Vgl. W. Bausch und H. Gohs: Kristallstruktur und Dauerbruch. Z. VDI **75**, 1935, pp. 1,450/51; F. Heber: Das Verhalten metallischer Werkstoffe im Bereich kleiner Verformungen. Report 5. Int. Congr. Appl. Mech., Cambridge, Mass., 1938.

⁴ H. F. Moore and J. R. Komers: An Investigation of the Fatigue of Metals. Univ. Illinois Bull. Engng. Exp. Station, **124**, 1921, p. 134.

⁵ H. F. Moore: How and when does a Fatigue Crack start? Metals and Alloys **7**, 1930, p. 297/99.

⁶ R. Ludwik: Die Bedeutung des Gleit- und Reistwiderstandes für die Werkstoffprüfung. Z. VDI **71** (1927), pp. 1532/38.

⁷ S. Laute und G. Sachs: Was ist Ermüdung? Mittsch. Mat.-Prüf.-Anst. **6**, 1929, pp. 89/91; Z. VDI **72**, 1928, pp. 1188/89.

⁸ N. Dawidow und E. Schewandin: Ueber den Ermüdungsriß. Metallwirtsch. **10**, 1931, pp. 710/14.

⁹ W. Kuntze: Zur Problemstellung der Metallermüdung. Metallwirtsch. **10**, 1931, pp. 65/97; Z. VDI **75**, 1931, pp. 285, and Mittsch. Mat.-Prüf.-Anst. **17**, 1931, pp. 48/52.

⁹ F. Bacon: Cracking and Fracture in Rotary Bending Tests. Engineering **134**, 1932, pp. 372/76; Fracture by Fatigue The Engineer **157** 1934, pp. 634/36.

¹⁰ F. Bacon: Fatigue and Corrosion Fatigue with Special Reference to Service Breakage. Proc. Instn. Mech. Engrs., London, **124**, 1933, pp. 685/736.

axles using bending alternating stress and applying the oil-whiting method. The free end of the test-piece was carried by a ball-bearing, but bent outwards by means of a spring; consequently, the highest strain took place where the groove turned to the increased section of the clamped part. The specimens were of 1 in. diameter; three test-pieces each of two different steels were loaded only a few tons/sq. in. above the fatigue strength. After cracks of about 0.1 in. length had been obtained, the specimens were subjected to loads of 90 (75), 65 and 50% of the fatigue strength. In the last case the cracks did not show further development. With increasing load, however, there was a reduction in the number of reversals between the starting of the crack and the final fracture. Further, the length of the crack increases slowly at the beginning with increase in the number of reversals, but becomes quicker as the final fracture is approached.

The number of reversals between the appearance of the first crack and the final fracture was considerably less with test pieces of 0.3 in. than of 1 in. diameter—i.e., the damage to the material was speeded up for equal load in specimens of smaller section; but the conditions of the tests were not equal; the small test pieces being strained by constant bending moments, whereas the larger specimens were obtained by bending moments which increased. Other investigations by the same worker on test pieces of 1 in. diameter showed a definite increase in the number of reversals with decreasing bending load.

Another report deals with big T specimens of gray-iron and white malleable cast-iron. The bending test was carried out with 1,000 reversals/min. and with constant deformation of the test pieces; the decrease of the number of oscillations, while the deflection remained constant, was considered as an indication of the beginning of the crack. The results confirmed generally those of the previous investigation.

Specimens of low-carbon steel of $\frac{1}{2}$ in. diameter and of different surface condition were investigated by the magnetic power method determining the length of the crack from 0.04 in. and 0.1 in. to the final fracture, which occurred when the crack had a length of about half the circumference of the test piece. The influence of rough and fine transversal or longitudinal scratches in the surface was unimportant. Again, it could be stated that the number of reversals between beginning of the crack and final fracture increased with decreasing load, but this relation was not so outstanding as in the previous investigations.

As may be seen from this summary of the results of existing reports, many attempts have been made to solve the numerous and complicated problems connected with fatigue fractures; but many contradictions and obscurities remain which need intensive and careful investigation. The object of the present work reported by Hempel is to clear the problem of the spreading of cracks; by changing the method of testing it should be possible to determine the form of test piece, the quantity of the load and the kind of material have on the spreading of a fatigue crack, when exposed to bending stress, and how far the contradictions as to the moment of starting of fatigue cracks could be modified. For this purpose, the magnetic powder method of investigation has been chosen making the cracks visible as quickly as possible. Simultaneously, it was hoped that recent investigations on the deformations of crystals at fatigue fractures would be facilitated.

The tests have been carried out on solid test pieces and on pieces with transversal holes. These were subjected to alternating bending stress in which the bending moment remained equal along the whole length of the specimen. Preliminary experiments are illustrated by magnetic powder pictures taken immediately after the starting of the crack, and the numerous stress reversals between the starting of the crack and the final fracture are given.

In the main experiments two steels are tested: one containing 0.05% C; 0.018% Si; 0.33% Mn; 0.02% P and 0.028% S; the other, 0.58% C; 0.18% Si; 0.72% Mn; 0.025% P and 0.050% S. The specimens were of 0.3 in. diameter, and the transversal holes in the drilled specimens were 0.043 in. diameter. The tensile strength of the two steels was 22.6 and 48.0 tons/sq. in., respectively.

In order to obtain information on the depth of the cracks and on the fatigue fracture surface, several series of experiments were carried out: a number of specimens were subjected each time to the same load, but with various running-times after the starting of the crack; thus, different crack depths should be obtained. The first very small traces of cracks could not be found by the magnetic powder method. As a measure of the spreading of the fatigue crack the number of stress reversals was used from the moment the crack could be identified to the moment when the experiment was stopped. This number of reversals may be termed the crack reversals, while those necessary for final fracture may be called fracture reversals.

As far as solid specimens were concerned, the crack reversals increased more or less quickly up to a crack depth of about 0.04 in., according to the load above the fatigue strength, but this increase diminished with increasing crack depth; as soon as the middle of the section is approached, the process of final fracture starts, and is finished after a few hundred reversals.

The rate of the crack spreading increased about proportionally to the square of the crack depth. This depth, as well as the surface of the fatigue fracture, has close relation to the crack length measured on the circumference of the specimen. In principle the process is the same with drilled test pieces, but the numbers of the crack reversals are different, i.e., the crack appears comparatively early, but the spreading takes place slower than with solid specimens.

In order to compare the results of both kinds of test pieces the ratios of crack reversals to fracture reversals and of the fatigue fracture surfaces to the sections of the specimens were calculated and plotted against each other. These diagrams show that, under similar loads the form of the specimen is of much greater influence than the load and the material. With solid test pieces the starting of the crack can be perceived at 85% to 90%, and with drilled test pieces at 30% to 60% of the fracture reversals.

Finally, a comparison is made of the results as given in other reports and those in the present investigations, especially with regard to the beginning of a fatigue crack. It could be confirmed that when alternating bending stress with equal bending moment over the measuring length is applied, a fatigue crack of solid test pieces takes place, almost independently of the test-piece diameter and of the material, after 70 to 90% of the fracture reversals.

Institute of Marine Engineers

THE following candidates have passed in all subjects of the May examinations for admission to Studentship of the Institute of Marine Engineers: E. Henshall, J. Lowson, J. M. Pratt (Rutherford Technical College, Newcastle); A. E. Richards, E. W. Youldon, J. B. Bremner, E. Grenfell, C. A. Rees (Falmouth T.S.); R. Bickenson (H.M.S. *Repulse*); J. H. Maycock (Erith T.C.); S. Hutchinson, B. Hildrew (Sunderland Technical College). B. Hildrew gained highest marks in four of the six subjects of the examination and the highest aggregate marks.

The following candidates qualified for Probationer Studentship: R. D. Hewitt and R. J. H. Stewart (Dover). Bryan Hildrew, on his results from this Studentship Examination, has been awarded the Lloyd's Register Scholarship, which is awarded annually by Lloyd's Register of Shipping. The Scholarship is valued at £100 a year for three years, and will enable this student to begin full-time study for the Honours Degree in Engineering of London University.

The Creep Strength of 17 Low-Alloy Steels at 1000° F.

The results of an investigation are given, the object of which was to determine the effect of composition and heat-treatment on the creep strength of the common low-alloy steels at 1,000° F. It is shown that the addition of chromium tends to lower the creep strength of carbon-molybdenum steel, but this decrease may be partially offset by increasing the molybdenum content. The stability of carbon-molybdenum steels is increased by the addition of chromium. Titanium or columbium improves the creep strength of chromium-molybdenum steels, but silicon and aluminium affect their creep strength adversely.

A WIDE variety of steels has been investigated with the object of determining the effect of composition and heat-treatment on the creep strength of the common low-alloys steels at 1,000° F. The results of this investigation are given in a paper by R. F. Miller, W. G. Benz and W. E. Unverzagt, which was presented at the recent annual meeting of the American Society for Testing Materials, and it is noteworthy that they show close agreement with the data published during the past year. A temperature of 1,000° F. was chosen for the comparison, and the same two stresses were used for all steels. The stresses, 10,000 and 15,000 lb. per sq. in., were selected because the available data indicated that they approximated the creep strength (stress corresponding to a creep rate of 0.000001 in. per inch per hour, or 1 % per 10,000 hours) of these steels.

Composition of Steels

The composition of the 17 steels used in this investigation is given in Table I. For comparative purposes, the steels have been divided into three groups.

- A: Carbon-molybdenum steels, with varying molybdenum and manganese contents.
- B: 1.25 to 3.25% chromium steels, with additions of molybdenum, silicon and aluminium.
- C: 5.0% chromium steels, with additions of molybdenum, silicon, aluminium, titanium and columbium.

TABLE I.—COMPOSITION OF STEEL.

| Steel. | Carbon, % | Manga- nese, % | Phos- phorus, % | Sulphur, % | Silicon, % | Chro- mium, % | Molyb- denum, % | Other Elements, % |
|----------------|--------------|----------------------|-----------------------|---------------|---------------|---------------------|-----------------------|-------------------------|
| <i>Group A</i> | | | | | | | | |
| A1 ... | 0.15 | 0.46 | 0.006 | 0.023 | 0.14 | — | 0.56 | — |
| A2 ... | 0.16 | 0.43 | 0.013 | 0.013 | 0.16 | — | 1.10 | — |
| A3 ... | 0.21 | 0.38 | 0.017 | 0.025 | 0.03 | — | 1.36 | — |
| A4 ... | 0.13 | 0.33 | 0.14 | 0.022 | 0.01 | — | 1.95 | — |
| A5 ... | 0.21 | 0.86 | 0.016 | 0.026 | 0.13 | — | 0.58 | — |
| <i>Group B</i> | | | | | | | | |
| B1 ... | 0.12 | 0.47 | 0.008 | 0.007 | 0.22 | 2.06 | 0.50 | — |
| B2 ... | 0.10 | 0.34 | 0.012 | 0.007 | 0.12 | 2.26 | 1.06 | — |
| B3 ... | 0.14 | 0.43 | 0.011 | 0.008 | 0.84 | 1.25 | 0.54 | — |
| B4 ... | 0.17 | 0.32 | 0.012 | 0.014 | 0.57 | 1.78 | 0.78 | — |
| B5 ... | 0.12 | 0.48 | 0.010 | 0.011 | 1.57 | 3.26 | 0.54 | — |
| B6 ... | 0.08 | 0.32 | 0.010 | 0.011 | 1.45 | 3.02 | 0.54 | Al 0.56 |
| <i>Group C</i> | | | | | | | | |
| C1 ... | 0.13 | 0.36 | 0.017 | 0.007 | 0.44 | 5.48 | 0.50 | — |
| C2 ... | 0.13 | 0.36 | 0.008 | 0.017 | 0.38 | 5.66 | 0.98 | — |
| C3 ... | 0.13 | 0.26 | 0.011 | 0.010 | 1.37 | 5.24 | 0.50 | — |
| C4 ... | 0.07 | 0.28 | 0.010 | 0.017 | 1.28 | 4.96 | 0.54 | Al 0.58 |
| C5 ... | 0.05 | 0.34 | 0.010 | 0.008 | 0.39 | 5.20 | 0.58 | Ti 0.47 |
| C6 ... | 0.09 | 0.42 | 0.012 | 0.011 | 0.35 | 5.00 | 0.55 | Cr 0.59 |

Heat-Treatment

The heat-treatment for each of these steels is shown in Table II. Suitable heat-treatments, representative of commercial practice, were selected. The material was heat-treated in the form of 1-in. rounds; after heat-treatment, coupons were cut from the ends of 8-in. bars for determination of micro-structure, and the remainder was machined to form the creep specimen. Two separate heat-treatments were employed for group A and for certain of the steels of groups B and C, in order to study further their effect on creep strength.

For the purpose of facilitating correlation of the data, the grain size of the steels was observed, and the results are given in Table III. The ferrite grains are those existing

TABLE II.—HEAT TREATMENTS.

| Steel. | Specimen. | Heat Treatment. |
|----------------|------------|--|
| <i>Group A</i> | | |
| A1-A4 | a and b | Normalise 1,650° F., 0.5 hour, air cool. |
| A1-A4 | c and d | Normalise 1,650° F., 0.5 hour, air cool; temper 1,300° F., 5 hours, air cool. |
| A5 | a and b | Normalise 1,650° F., 0.5 hours, air cool; temper 1,300° F., 5 hours, air cool. |
| <i>Group B</i> | | |
| B1 | a and b | Normalise 1,650° F., 1 hour, air cool; temper 1,380° F., 4 hours, furnace cool. |
| B2 | a and b | Normalise 1,650° F., 1 hour, air cool; temper 1,380° F., 4 hours, furnace cool. |
| B3 | a | Anneal 1,550° F., 1 hour, furnace cool. |
| B3 | b, c and d | Normalise 1,650° F., 1 hour, air cool; temper 1,380° F., 4 hours furnace cool. |
| B4 | a and b | Normalise 1,750° F., 1.5 hours, air cool; temper 1,375° F., 2 hours, furnace cool. |
| B5 | a, b and c | Anneal 1,600° F., 2 hours, furnace cool. |
| B6 | a, b and c | Normalise 1,550° F., 3 hours, air cool. |
| <i>Group C</i> | | |
| C1 | a and b | Anneal 1,600° F., 2 hours, furnace cool. |
| C1 | c, d and e | Normalise 1,600° F., 1 hour, air cool; temper 1,380° F., 4 hours, air cool. |
| C2 | a and b | Normalise 1,650° F., 0.5 hour, air cool; temper 1,380° F., 4 hours, air cool. |
| C3 | a, b and c | Anneal 1,600° F., 2 hours, furnace cool. |
| C4 | a, b and c | Normalise 1,550° F., 3 hours, air cool. |
| C5 | a and b | Temper 1,380° F., 4 hours, air cool. |
| C6 | a | Anneal 1,600° F., 1 hour, furnace cool. |
| C6 | b, c and d | Temper 1,380° F., 4 hours, air cool. |

in the steel at room temperature after the finish heat-treatment and before creep testing, determined from the coupons cut from the heat-treated bars before machining. All the steels are fine grained in the ferritic condition. The austenite grains are those produced at the maximum heat-treating temperature for each steel.

Creep Strength at 1,000° F.

The creep-testing apparatus employed in these tests has been described by Miller, Campbell, Aborn and Wright,¹ and by Miller.² During the course of this testing programme

TABLE III.—GRAIN SIZE.

| Code. | Heat Treatment, deg. Fahr. | Grain Size Number, ASTM. | | |
|-------|----------------------------|--------------------------|----------------------------|--------------|
| | | Ferrite. | Austenite. | McQuaid-Ehn. |
| A1 | 1,650 N | 8-9 | 7-8 | 6-7 |
| A2 | 1,650 N 1,300 T. | 8-9 | 7-8 | 6-7 |
| A3 | 1,650 N | 8-9 | 7-9 | 8-9 |
| A4 | 1,650 N 1,300 T | 9 | Below A ₃ temp. | 2-3; 6-7 |
| A5 | 1,650 N | 9 | Below A ₃ temp. | 2-3; 6-7 |
| B1 | 1,650 N 1,380 T | 8-9 | Below A ₃ temp. | 8-9 |
| B2 | 1,650 N 1,380 T | 8-9 | Below A ₃ temp. | 8-9 |
| B3 | 1,650 An | 8-9 | Below A ₃ temp. | 8-9 |
| B4 | 1,750 N 1,375 T | 8-9 | Below A ₃ temp. | 8-9 |
| B5 | 1,600 An | 8-9 | Below A ₃ temp. | 8-9 |
| B6 | 1,550 N | 8-9 | Below A ₃ temp. | 8-9 |
| C1 | 1,600 An | 8-9 | Below A ₃ temp. | 8-9 |
| C2 | 1,600 N 1,380 T | 9-11 | Below A ₃ temp. | 8-9 |
| C3 | 1,650 N 1,380 T | 9-11 | Below A ₃ temp. | 8-9 |
| C4 | 1,600 An | 10-11 | Below A ₃ temp. | 8-9 |
| C5 | 1,650 N | 7-8 | Below A ₃ temp. | 8-9 |
| C6 | Hot rolled and 1,380 T | 8-9 | Below A ₃ temp. | 8-9 |
| C6 | Hot rolled and 1,380 T | 7-8 | Below A ₃ temp. | 8-9 |
| C6 | 1,600 An | 11-12 | Below A ₃ temp. | 8-9 |

* The following terms and abbreviations are employed:
N = normalise—air cool from top heat-treating temperature.
An = anneal—furnace cool from top heat-treating temperature.
T = temper—reheat below lower critical temperature.

¹ R. F. Miller, R. F. Campbell, B. H. Aborn and E. C. Wright, "Influence of Heat Treatment on Creep of Carbon-Molybdenum and Chromium-Molybdenum-Silicon Steels," *Transactions, Am. Soc. Metals*, Vol. 26, p. 81 (1935).

² Richard F. Miller, "The Strength of Metals at Elevated Temperatures," *Mechanical Engineering*, vol. 61, No. 8, August, 1939, p. 589.

an improved creep-measuring microscope was developed and installed, which has a sensitivity of ± 0.00002 in. and a maximum precision of ± 0.00001 in. per inch. The specimens have a diameter of 0.505 in. and a gauge length of 2.3 in. The temperature variation over the gauge length of the specimen is $\pm 1^\circ$ F., and the control is $\pm 1.5^\circ$ F. The irregularities in the extension measurements, and hence in the creep curves, were of about the same order of magnitude as the changes in length produced by the $\pm 1.5^\circ$ F. fluctuation of the temperature in the furnace. All creep tests were carried out in conformity with the A.S.T.M. Tentative Method of Test for Long-Time (Creep) High Temperature Tension Tests of Metallic Materials (E22-38T).³

The stress employed for each individual specimen, the duration of the test, and the creep rate during the last 1,000 hours of each test are listed in Table V. From these data, the creep strength of each steel has been calculated, as is shown in Table V.

The creep strength of the carbon-molybdenum steels increases with increasing molybdenum content, both in the normalised and in the normalised-and-tempered condition, as shown in Fig. 1. For each composition, the creep strength of the normalised material is higher than that of the normalised-and-tempered material. However, the normalised material is less stable than the normalised-and-tempered material, and the creep strength of both may become the same after a prolonged exposure to a temperature of $1,000^\circ$ F. Increase of manganese from 0.46 to 0.86% causes a slight increase in the creep strength of normalised-and-tempered carbon 0.5% molybdenum steel, as seen from a comparison of steels A1 and A5, Table V.

Fig. 2 shows graphically the effect on creep strength of change of composition of similarly heat-treated chromium-molybdenum steels. The creep strength of normalised-and-tempered steel containing 2% or 5% chromium is enhanced by raising the molybdenum content from 0.5% to 1%; but it is lowered by raising the chromium from 2% to 5%. A similar lowering is observed when the chromium content of normalised chromium-molybdenum-silicon-aluminum steel is raised from 3% to 5%. On the other hand, increase of chromium from 3% to 5% resulted in little change of the creep strength of annealed steel containing 0.5% molybdenum and 1.5% silicon.

Steels B3 and B4 both contain chromium, molybdenum, and silicon, were similarly heat-treated, and have a similar grain size. The slightly higher alloy content of B4 results in a considerably more creep-resistant material. A possible reason for this enhanced creep strength is that this small change in composition markedly increases the air-hardening propensities of this type of material; in other words, the same normalising and tempering heat-treatment yields a different microstructure in the two steels. In steel B4, the carbide particles are found to be fine and uniformly distributed, while in steel B3 they have coalesced at the grain boundaries where they are presumably less effective in strengthening the material.

The creep strength of steel B3 was found to be 7,800 lb. per sq. in. at $1,000^\circ$ F., considerably lower than that, 22,000 lb. per sq. in., reported by another laboratory. Consequently, a sample of steel B3 was heat-treated

(annealed) in the manner reported to have been employed in obtaining the higher creep strength, and this sample was tested at a stress of 22,000 lb. per sq. in. The creep rate during the last 1,000 hours of this test was 0.000024 in. per inch per hour, a rate so rapid that the lower value was confirmed as being a safer estimate of its creep strength.

Remarkably good results were obtained with steels C5 and C6 in the hot-rolled and tempered condition. Steel C5 shows a creep strength of 11,500 lb. per sq. in., as compared to 6,600 lb. per sq. in. for the same steel (C1) in the normalised-and-tempered condition without titanium. Steel C6 has a creep strength of 16,400 lb. per sq. in., the best of any steel tested, except the normalised-carbon-molybdenum steel A4, which has a creep strength of 18,800 lb. per sq. in. The creep rate under a stress of 15,000 lb. per sq. in. at $1,000^\circ$ F., of steel C6 in the annealed condition was observed to be 0.000042 in. per inch per hour, but in the hot-rolled and tempered condition to be only 0.0000028 in. per inch per hour; this is about 150 times

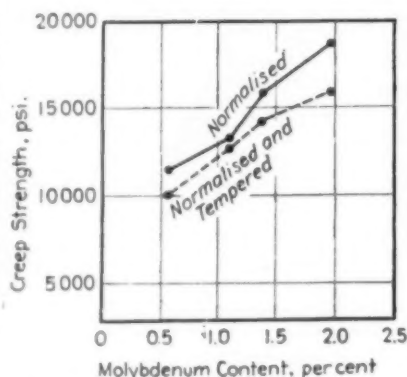


Fig. 1.—Effect of molybdenum content on creep strength of carbon-molybdenum steels at $1,000^\circ$ F. (Group A).

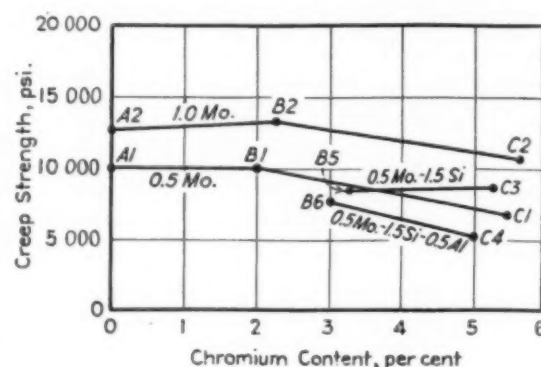


Fig. 2.—Effect of chromium content on creep strength of chromium-molybdenum steels at $1,000^\circ$ F. (Groups B and C).

less rapid. Although the creep strength of steel C6 is lower in the annealed than in the hot-rolled and tempered condition, it is still above that of the same steel without columbium. This is demonstrated by a comparison of the creep rates of steels C1 (without columbium), and C6 (with columbium) in the annealed condition under a stress of 15,000 lb. per sq. in. (0.000094 and 0.000042 in. per inch per hour, respectively).

In general, it may be said that, excepting the titanium- and columbium-bearing steels, the creep strength of the steels in group C is somewhat lower than that of the steels in group B, while that of group B is somewhat lower than that of group A.

Stability of Microstructure

The microstructure of all the specimens was examined, both before and after creep testing, and the observations regarding the relative stability of the microstructures of these steels at $1,000^\circ$ F. are summarised in Fig. 3. The change of microstructure observed in a given steel with a given heat-treatment seemed to be unaffected by the stress, and to depend solely on the duration of its sojourn at $1,000^\circ$ F.

The most pronounced change of microstructure was found in the normalised carbon-molybdenum steels of group A. There was a moderate change in the normalised-and-tempered carbon-molybdenum steels, and in the annealed steel B3. In the other chromium-molybdenum steels of groups B and C there was a very slight change, or no visible change at all. It is thought that the superior stability of the steels of groups B and C, as compared to those in group A may be ascribed to the chromium content, because both in the pearlitic and in the spheroidized condition, additions of chromium increased the stability of the microstructure.

Structures that are unstable at $1,000^\circ$ F. may be so altered

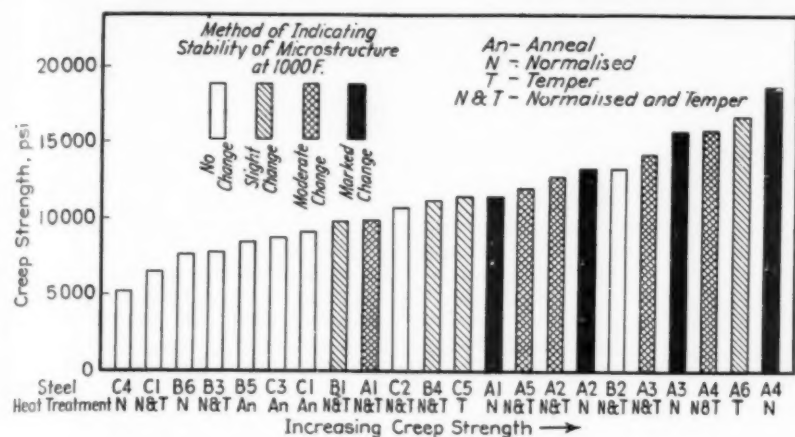


Fig. 3.—Steels plotted in order of increasing creep strength, showing relative stability of micro-structure during creep test at 1,000°F.

TABLE IV.—TEST DATA, 1,000°F. CREEP TESTS.

| Type of steel. | Specimen. | Creep Stress, lb./sq.in. | Duration of Creep. | Creep Rate during last 1,000 hrs. 0.000001 in. per in. per hour. | Brinell Hardness. | | Modified Charpy Impact Strength, ft.-lb.* | |
|----------------|-----------|--------------------------|--------------------|--|--------------------|-------------------|---|-------------------|
| | | | | | Before Creep Test. | After Creep Test. | Before Creep Test. | After Creep Test. |
| | | | | | | | | |
| Group A. | | | | | | | | |
| A1 | a | 15,000 | 5,000 | 5-16 | 131 | 134 | 39 | 34 |
| | b | 10,000 | 3,200 | 0-52 | 131 | 134 | 39 | 38-5 |
| | c | 15,000 | 3,100 | 17-1 | 116 | 128 | 40 | 37 |
| | d | 10,000 | 3,000 | 1-02 | 116 | 114 | 40 | 40 |
| A2 | a | 15,000 | 4,150 | 1-65 | 146 | 179 | 39 | 33-5 |
| | b | 10,000 | 3,100 | 0-43 | 146 | 174 | 39 | 38-5 |
| | c | 15,000 | 2,900 | 2-13 | 143 | 137 | 42 | 42 |
| | d | 10,000 | 3,300 | 0-39 | 143 | 140 | 42 | 41 |
| A3 | a | 15,000 | 3,000 | 0-88 | 167 | 197 | 26-5 | 24-5 |
| | b | 10,000 | 3,000 | 0-36 | 167 | 201 | 26-5 | 25 |
| | c | 15,000 | 3,000 | 1-17 | 163 | 163 | 29 | 31-5 |
| | d | 10,000 | 2,000 | 0-52 | 163 | 174 | 29 | 32-5 |
| A4 | a | 15,000 | 3,500 | 0-47 | 152 | 212 | 42 | 18-5 |
| | b | 10,000 | 3,000 | 0-17 | 152 | 223 | 42 | 6 |
| | c | 15,000 | 2,000 | 0-78 | 173 | 167 | 38 | 38-5 |
| | d | 10,000 | 2,600 | 0-22 | 173 | 163 | 38 | 39 |
| A5 | a | 15,000 | 3,000 | 2-34 | 152 | 156 | 38 | 36 |
| | b | 10,000 | 2,400 | 0-56 | 152 | 152 | 38 | 33 |
| Group B. | | | | | | | | |
| B1 | a | 15,000 | 1,900 | 8-8 | 141 | 131 | 49 | 45 |
| | b | 10,000 | 3,000 | 1-04 | 141 | 134 | 49 | 45 |
| B2 | a | 15,000 | 3,000 | 1-47 | 159 | 167 | 46 | 44 |
| | b | 10,000 | 3,000 | 0-54 | 159 | 163 | 46 | 42-5 |
| B3 | a | 22,000 | 2,100 | 24-2 | 156 | 159 | 41-5 | 21 |
| | b | 15,000 | 1,300 | 33-6 | 146 | — | 36 | — |
| | c | 10,000 | 3,000 | 3-8 | 146 | 152 | 36 | 33 |
| | d | 7,000 | 2,300 | 0-59 | 146 | 137 | 36 | 35 |
| B4 | a | 15,000 | 3,000 | 2-9 | 181 | 163 | 40-5 | 27 |
| | b | 10,000 | 3,000 | 0-71 | 181 | 179 | 40-5 | 26 |
| B5 | a | 15,000 | 800 ^b | 87-6 | 170 | 170 | 31 | 26 |
| | b | 10,000 | 3,000 | 4-12 | 170 | 170 | 31 | 20 |
| | c | 7,500 | 3,000 | 0-67 | 170 | 167 | 31 | 16-5 |
| B6 | a | 15,000 | 200 ^b | 98-1 | 163 | 159 | 21 (2) | 27-5 |
| | b | 10,000 | 1,000 ^b | 5-7 | 163 | 170 | 21 (2) | 26-5 |
| | c | 6,500 | 3,000 | 0-4 | 163 | 156 | 21 (2) | 25-5 (1-5) |
| Group C. | | | | | | | | |
| C1 | a | 15,000 | 200 ^b | 94-0 | 136 | 134 | 38 | 31-5 |
| | b | 10,000 | 2,000 | 2-0 | 136 | 146 | 38 | 31-5 |
| | c | 15,000 | 1,000 ^b | 13-9 | 191 | 174 | 50 | 45 |
| | d | 10,000 | 3,000 | 3-56 | 191 | 172 | 50 | 43 |
| | e | 7,000 | 3,000 | 1-19 | 191 | 179 | 50 | 43 |
| C2 | a | 15,000 | 1,000 ^b | 7-9 | 202 | 174 | 39 | 33 |
| | b | 10,000 | 2,200 | 0-72 | 202 | 212 | 39 | 35 |
| C3 | a | 15,000 | 500 ^b | 84-3 | 170 | 170 | 38 | 35-5 |
| | b | 10,000 | 2,200 | 4-56 | 170 | 167 | 38 | 23-5 |
| | c | 7,500 | 3,000 | 0-30 | 170 | 170 | 38 | 28 |
| C4 | a | 15,000 | 200 ^b | 163 | 154 | 163 | 51 | 46-5 |
| | b | 10,000 | 1,700 | 18-8 | 154 | 149 | 51 | 40 (1-5) |
| | c | 5,500 | 2,200 | 1-2 | 154 | 152 | 51 | 41 |
| C5 | a | 15,000 | 1,500 | 32-3 | 149 | 134 | 52 | 44 |
| | b | 10,000 | 2,200 | 0-2 | 149 | 126 | 52 | 43 |
| C6 | a | 15,000 | 2,000 | 0-28 | 169 | 156 | 49 | 36 |
| | b | 10,000 | 2,000 | 0-28 | 169 | 149 | 49 | 43 |
| | c | 20,000 | 1,500 | 12-8 | 169 | 163 | 49 | 47-5 |
| | d | 15,000 | 900 | 41-7 | 140 | 140 | 46-5 | 38 |

* Values are averages of two tests. When results are in wide disagreement the second test value is indicated in parenthesis.

^b The creep rate exceeded the capacity of the apparatus; other tests were run for longer times under lower stresses to supply data necessary for determination of stress for desired creep rate.

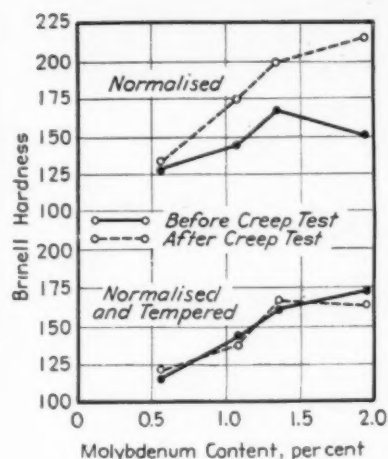


Fig. 4.—Effect of molybdenum content on hardness of carbon molybdenum steels before and after creep testing at 1,000°F. (Group A).

by heating for a few hours at about 1,300°F. that they will remain essentially unchanged, however long they are exposed to the lower temperature of 1,000°F. In other words, the same amount of structural change may be brought about in a few hours at the higher temperature that would have taken thousands of hours at the lower testing temperature.

Examination of Fig. 3 shows that, in general, the steels with the highest creep strength are the ones with the least stability. Also, that when two heat-treatments are used for the same steel, the more stable structure has the lower creep strength. If the creep tests had been run long enough at 1,000°F. to permit the same change of structure which was produced by the subcritical heating (tempering) before creep testing, it is probable that the high creep strength data would not have been obtained. In this case, even 3,000 hours was not long enough to disclose the magnitude of the decrease of creep strength which would be encountered.

The design engineer would be well advised to inquire whether the microstructure of a steel he is considering for a given application is stable at the operating temperature, since the creep strength quoted for a given steel may represent a transitory condition and over a period of time the creep strength of the material may decrease considerably. Conservative estimates of the creep strength of a given material should be made on the basis of a structure known to be stable at the service temperature.

Impact Strength and Hardness

The impact strength of the steels, as determined before and after creep testing, is given in Table IV. In order to determine the impact strength of the steels before creep testing, specimens of each steel 1 in. in diameter by 4 in. long were treated in the same manner as the creep specimens, and two duplicate Charpy keyhole impact specimens were then machined from each bar. Since the creep specimens from which the keyhole Charpy impact specimens were machined after creep testing had a diameter of 0.0505 in., it was necessary to employ the sub-standard two-thirds-size specimen (hereafter called the "modified" Charpy impact specimen), which has a cross-section measuring 6 mm. by 10 mm., instead of 10 mm. by 10 mm.

The only significant effect on impact strength which could be attributed to composition occurred in steels B6 and C4. The initial heat-treatment of these materials resulted in exceedingly coarse ferrite grains (A.D.T.M. No. 0 to 2) in the outer portions of the bar stock. Even though an attempt was made to machine off this coarse-grained skin in the preparation of the creep specimens, it is believed that some of this coarse-grained material

TABLE V.—CREEP STRENGTH AT 1,000° F.

| Code. | Heat Treatment, deg. Fahr. | Stress for Creep Rate of 0.00001 in. per inch per hour, lbs./sq. in. | |
|----------|-------------------------------|---|---------------------|
| | | Kearny. | Other Laboratories. |
| A1 | 1,650 N | 11,500 | { 12,400 (B & W) |
| | 1,650 N 1,300 T | 10,000 | { 10,800 (Timken) |
| A2 | 1,650 N | 13,300 | { 11,600 (B & W) |
| | 1,650 N 1,300 T | 12,800 | |
| A3 | 1,650 N | 15,800 | |
| | 1,650 N 1,300 T | 14,300 | |
| A4 | 1,650 N | 18,800 | |
| | 1,650 N 1,300 T | (extrapolated) | |
| | 1,650 N 1,300 T | 15,900 | |
| A5 | 1,650 N 1,300 T | 12,000 | |
| B1 | 1,650 N 1,380 T | 10,000 | { 8,900 (Timken) |
| | | | { 11,000 (B & W) |
| B2 | 1,650 N 1,380 T | 13,300* | { 11,000 (Timken) |
| | | | { 16,350 (B & W) |
| B3 | 1,650 N 1,380 T | 7,800* | { 22,000 (Timken) |
| B4 | 1,750 N 1,375 T | 11,300 | |
| B5 | 1,000 An | 8,500* | |
| B6 | 1,550 N | 7,800* | { 8,700 (Timken) |
| | | | { 8,400 (Timken) |
| C1 | 1,000 An | 9,100* | { 9,200 (B & W) |
| | | | { 10,100 (Timken) |
| | | | { 9,725 (Climax) |
| C2 | 1,600 N 1,380 T | 6,600* | |
| | 1,650 N 1,380 T | 10,800* | |
| C3 | 1,400 An | 8,800* | { 10,000 (Climax) |
| C4 | 1,550 N | 5,200* | { 8,700 (Timken) |
| | | | { 7,600 (Timken) |
| C5 | 1,380 T | 11,500 | { 13,000 (B & W) |
| C6 | 1,780 T | 16,400 | |

* Stable structure.

was included in two of the impact specimens of steel B6 and in one of the impact specimens of steel C4, resulting in the abnormally low modified Charpy impact values of 1.5 and 2.0 ft.-lb. No consistent trend in impact values could be attributed to differences in the heat-treatment of individual steels.

Exposure of the steels to a temperature of 1,000° F. for 3,000 hours caused little if any change in the impact strength of the carbon-molybdenum steels in group A.

Steel A4 was an exception; in the normalised condition, the modified Charpy impact strength decreased from 42.0 ft.-lb. in the initial material to 18.5 and 6.0 ft.-lb. after creep testing under 15,000 and 10,000 lb. per sq. in., respectively. This may be associated with the corresponding increase in hardness from 152 Brinell to 212 and 223 Brinell. The chromium-molybdenum steels in groups B and C show a tendency toward lower impact strength after creep testing, and the decrease is more pronounced in the steels containing silicon.

The hardness of the steels was determined before creep testing on the micro-structure coupons cut from the heat-treated bars before machining the creep specimens. The hardness of the specimens after creep testing was measured on the broken impact specimens which were machined from the creep specimens. The values are given in Table IV. The hardness of the normalised carbon-molybdenum steels (group A) increases with increasing molybdenum content, except in steel A4, which contains only 0.13% carbon. A marked increase in the hardness of these normalised carbon-molybdenum steels occurs during creep testing as shown graphically in Fig. 4. It will also be seen that the hardness of the normalised and tempered carbon-molybdenum steels increases with increasing molybdenum content, but that in this condition there is very little change after creep testing.

In the chromium-molybdenum steels, the only significant trend attributable to change of composition is that an increase of molybdenum content in steels containing 2% or 5% chromium results in an increase in hardness, both before and after creep testing.

After exposure to the creep test at 1,000° F., the chromium-molybdenum steels of groups B and C show either a slight decrease of hardness or no change.

Factors Influencing the Formation of a Smooth Coating of Tin on Copper by Hot Dipping

AN investigation of the factors which affect the angle of contact of dewetted tinned copper is described in a paper by Dr. Bruce Chambers and R. H. Wadie*. Particular attention is directed to such factors as the fineness of copper, the time and temperature of the dipping process, and the effect of the pressure of copper, lead, bismuth, antimony, zinc and iron in the tinbath. In order to make conditions so severe that dewetting took place under a wide range of conditions, resin was chosen as a flux and cover in the investigation.

The author points out that the amount of dewetting that takes place depends upon the quantity of tin present on the surface immediately before dewetting, and on the time during which the coating material is sufficiently fluid to flow. Thus, the amount of dewetting depends on a number of factors, some of which are not related to the dewetting tendency. The amount of dewetting, therefore, does not form a reliable guide to the extent of this tendency, which is revealed by the angle of contact. This angle may, therefore, be called the angle of dewetting, and a method is described for estimating the tendency to dewet by measurement of the angle of contact of the tin with the surface from which it has dewetted. On the basis of the results obtained by this investigation it is stated that suitable modifications of the conditions of dipping can be chosen so that the tendency to dewetting disappears.

No attempt has been made to imitate practical conditions either in regard to the preparation of the surface of the copper or in the choice of a flux. The choice of resin was governed, in fact, by the observations that comparatively large angles of dewetting could be obtained. This made it possible to vary the conditions considerably and still

obtain measurable dewetting. It is evident that the general principles that have emerged from this work will apply to any kind of hot dipping, and in particular to the tinning of iron and steel by this method. In this connection, two points may be mentioned: in the first place, the very wide difference in the tinning quality of various steels may be bound up with the value of the specific surface, as the presence of untinnable inclusions or constituents will reduce it below unity, the result being a higher angle of contact and a greater tendency to dewet. Secondly, the possibility of measuring the extent of dewetting by the method described in the paper may make it possible to decide what alteration in conditions is necessary to avoid dewetting in any actual case.

Corrosion of Magnesium Alloys

TESTS were carried out with aircraft alloys containing 5.8 to 5.72% aluminium, 2.78 to 2.81% manganese, 0.08 to 0.09% phosphorus and 0.07 to 0.08% silicon. No serious corrosion was observed after pickling for 10 mins. in 20% chromic acid solution which had a very good cleansing action. The magnesium alloys were dissolved at the rate of 0.025 to 0.035 mm. per min. when oxidised in aqueous solutions containing, respectively, 5% potassium bichromate, 11% nitric acid (d 1.4) and 0.1% ammonium chloride. Such baths required to be renewed after treatment of 0.08 sq. metre of alloy surface per litre. A bath recommended for oxidation of magnesium-alloy screws consisted of 10 grms (per litre) arsenious acid and 5 grms. sodium chloride. A. T. Bundin, *Aeronautics Industry* (U.S.S.R.), 1939, No. 9, p. 49-59.

*Jour. Inst. Metals, 68, pp. 241-252. (July, 1940).

The Constitution and Structure of Alloys of Intermediate Composition in the Systems Copper-Indium, Copper-Aluminium, Copper-Germanium Silver-Aluminium and Silver-Germanium

THE experimental work described in this paper¹ refers to the phase boundaries of the solid phases only, and may be summarised under the following headings:

1. The constitution of copper-indium alloys in the regions 18 to 25 atomic per cent. of indium.
2. The constitution of copper-aluminium alloys in the region of the β -eutectoid point.
3. The constitution of copper-germanium alloys in the region 10 to 20 atomic per cent. germanium.
4. The constitution of silver-aluminium alloys in the region 15 to 45 atomic per cent. of aluminium.
5. Experiments on the system silver-germanium.
6. Experiments on the crystal structure of 3/2 electron compounds.

This work forms part of a comprehensive study of the factors affecting the formation of 3/2 electron compounds in alloys of copper, silver, and gold with the elements of the β sub-groups. The authors hoped to determine the phase boundaries of the systems concerned in the ranges of composition affecting the 3/2 electron compounds of the silver and copper alloys. Circumstances have arisen, however, which make it clear that the complete experimental programme will not be finished for several years, and, to prevent undue accumulation of experimental work, those sections of the work which establish the phase boundaries and structures of some of the solid phases are published in the present paper.

The β -Phase of the System Copper-Indium

The equilibrium diagram of the system copper-indium has been investigated by Weibke and Eggers,² but it was decided to reinvestigate the β -phase area. The amount of indium available prevented an accurate determination of the liquidus curve, but solidus and solid-solubility curves of the β -phase have been determined accurately and also some points on the α solid-solubility curve.

The general result of the present determination is slightly to extend the β -phase area as compared with that given in the diagram of Weibke, but to show clearly that the phase does not extend quite up to the composition 25 atomic per cent. of indium, corresponding to the formula Cu_3In . It is also established that the eutectoid composition, although very near to that required by the formula Cu_4In , is slightly on the indium-rich side of this simple atomic ratio.

The β -Phase of the System Copper-Aluminium

The general form of the β -phase area in the equilibrium diagram of the system copper-aluminium was established by the work of Stockdale,³ but the exact composition and temperature of the eutectoid point have been in dispute. It is also known that in rapidly-cooled alloys the transformation of the β -phase to a mixture of the α and δ phases takes place through the formation of intermediate phases, but doubt has existed as to whether these phases are part of the equilibrium diagram of the stable system. Some of the discrepancies between the results of previous investigators are probably due to the varying purity of the aluminium used, and the authors have therefore re-investigated the equilibrium diagram at the extreme tip of the β -phase area.

The results show that the β -phase area narrows to a simple eutectoid point, and that under equilibrium conditions

there is no indication of the existence of a second modification of the β -phase. The hexagonal β' phase obtained by Obinata,⁴ and the phase with a structure resembling that of a face-centred cube, found by Ageew and Kurdjumow,¹⁰ Bradley and Jones¹¹ and Wassermann,¹² are both metastable intermediate forms, and do not belong to the stable diagram. The readiness with which the intermediate forms are produced suggests, however, that the characteristics of the system copper-aluminium are mid-way between those of the system copper-gallium in which the β phase changes into a stable close-packed hexagonal phase, and the system copper-indium in which the β phase breaks down directly into $\alpha + \delta$.

The eutectoid temperature lies between 560° and 565° C. in contradiction to the value 537° C. given by Stockdale. The present value is in fair agreement with the temperature (570° C.) given by Smith and Lindlie,¹³ and in good agreement with the values 566°, and 560° C. given by Curry¹⁴ and by Braesco,¹⁵ respectively. The eutectoid composition lies at 24.0 to 23.9 atomic per cent. of aluminium in good agreement with the work of Stockdale, and of Smith and Lindlie. The suggestion of Bradley and Jones that the eutectoid occurs at 12.4 weight per cent. aluminium (25 atomic per cent.) is not confirmed, and Stockdale's¹⁶ criticisms of the methods employed by these authors appear to be justified.

The ζ -Phase Region of the System Copper-Germanium

The α solid-solution area of the system copper-germanium was investigated by Hume-Rothery, Mabbott, and Channell-Evans,⁴ whilst the complete equilibrium diagram was studied by Schwarz and Elstner,¹⁷ and Maucher,¹⁸ who did not, however, carry out annealing experiments for the accurate determination of the phase boundaries. The rarity of germanium has prevented the authors from carrying out further experiments on the liquidus, but in view of the interest of the system for the theory of copper alloys, they have determined the phase boundaries of the ζ phase,³ and have also re-determined the α solid-solubility curve.

The present work shows that the ζ phase extends over a considerably wider area than was indicated by the diagrams of Schwarz and Elstner, and of Maucher. On the copper-rich side, the ζ -phase area extends very much nearer to the α -phase area than was previously supposed. The $\zeta/\zeta + \gamma$ boundary determined in the present work is distinctly on the germanium-rich side of that given by Schwarz and Elstner, but agrees approximately with that of Maucher, who did not, however, note the change in direction when the $\zeta + \epsilon$ region is reached. Although the quantity of germanium did not permit a detailed investigation, we conclude that the $\gamma \rightleftharpoons \epsilon$ transformation is at approximately 540° C., and is nearer to the value (538° C.) of Schwarz and Elstner than to that of Maucher (558° C.).

The Equilibrium Diagram of the System Silver-Aluminium in the Region 15-40 Atomic Per Cent Aluminium

The equilibrium diagram of the system silver-aluminium in the region 15-40 atomic per cent. of aluminium has been in considerable dispute. All of the more modern investigators agree that a typical β -phase exists at high temperatures, and, by work with a high-temperature X-ray camera,

the authors have confirmed that this phase has a body-centred cubic structure. This β phase is stable only above 603° C., whilst at low temperatures a μ -phase,³ with a β -manganese structure, exists over a narrow range of composition in the region of 25 atomic per cent. aluminium. The diagram of Crepaz¹⁹ is theoretically untenable, whilst according to Ageew and Shoyket,²⁰ the β phase decomposes into a eutectoid mixture of α and ϵ phases, the latter having a close-packed hexagonal structure. These phase boundaries do not agree with work of Hofmann and Volk,²¹ who showed that an alloy containing 25 atomic per cent. of aluminium had a body-centred cubic structure at 700° and 605° C., and a close-packed hexagonal structure at 570° and 465° C., whereas the diagrams of Ageew and Shoyket and of Obinata and Hagiya would require the structures at these last two temperatures to consist of a mixture of α and ϵ phases. The results of Hofmann and Volk would, however, be in agreement with the diagram of Ishida, Tajiri, and Karasawa.²⁶ In the present work the authors have reinvestigated this part of the diagram in great detail, and have explained some of the above discrepancies. The work also led to the discovery of a most remarkable volatilisation effect in these alloys.

When compared with the diagrams of the previous investigators mentioned, the results of the present work show that the close-packed hexagonal ϵ phase extends considerably further towards the silver-rich side of the diagram than had previously been supposed, and the work of Hofmann and Volk is confirmed. The diagram of Ishida, Tajiri and Karasawa²⁶ is clearly incorrect as regards the α solid-solubility curve, although their ϵ -phase boundaries do not differ by more than two atomic per cent. from the authors', except as regards the details at the temperatures of the $\beta \rightarrow \epsilon$ and the $\alpha + \epsilon \rightarrow \mu$ transformations. The narrowing of the ϵ -phase area at low temperatures found by these workers, and by Obinata and Hagiya,²⁷ is confirmed, and the longer times of annealing used in the present work show that it is more pronounced than had been imagined. The different temperatures found for the $\alpha + \epsilon \rightarrow \mu$ transformation may be connected with the varying purity of the aluminium used by different workers.

The present work serves to emphasise the interesting relations existing between the body-centred cubic β , and the close-packed hexagonal ϵ phase. In this system as in the systems copper-gallium and silver-zinc, the β - and ϵ -phase areas overlap as regards composition, whilst as the authors pointed out in the system copper-aluminium, a close-packed hexagonal structure is readily formed on quenching the body-centred cubic β alloys. There is thus a very close relation between the β and ϵ phases, and the present work shows, further, that the close-packed hexagonal ϵ phases of electron concentration 3/2 may join up continuously with the close-packed hexagonal ϵ phases of electron concentration 7/4, since the present close-packed hexagonal ϵ phase includes both electron concentrations. In the system silver-aluminium there is no phase with a " γ -brass" structure; apart from the existence of a μ phase, the main characteristic of the system is that the typical ϵ and ϵ' close-packed hexagonal phases, which usually occur at electron concentrations of 3/2 and 7/4, respectively, have united to form a single phase at the expense of the " γ -brass" type, which usually occurs at an electron concentration of 21/13.

Experiments on the System Silver-Germanium

The system silver-germanium has been studied by Briggs, McDuffie, and Willisford,²⁵ but their investigation was not very detailed. Alloys containing 10, 16.6, and 20 atomic per cent. of germanium were therefore prepared. The alloys were annealed at 586° C. for nine days, after which samples were examined microscopically and re-annealed at 298° C., 404° C., and 500° C. for periods of eight weeks, 24 days, and 15 days, respectively. The microstructures showed all the alloys to consist of the

primary α solid solution in silver, together with germanium, and there was no indication of the existence of a homogeneous phase in the region of a 3/2 electron compound. This conclusion has also been reached by Maucher,¹⁸ whose work was published after the above experiments were completed. The authors have redetermined the temperature of the $\alpha +$ germanium eutectic by very accurately-controlled quenching experiments, and have shown that this lies between the limits 650.5° and 651.6° C. The eutectic temperature is thus 651° C., and is very slightly higher than the values 650° and 649° C. given by Briggs, McDuffie, and Willisford, and by Maucher respectively.

1 By W. Hume-Rothery, G. V. Raynor, P. W. Reynolds, and H. K. Packer, *Jour. Inst. Metals*, 1940, **60**, pp. 209-239.

2 The authors have used the symbol ϵ because the phase has a close-packed hexagonal structure at an electron concentration of 1.5, the symbol ϵ' being used for the close-packed hexagonal structures at electron concentration 7/4.

3 In accordance with the notation of the previous investigation, the authors use the symbol μ for the β -manganese structure, and the symbol ϵ for the close-packed hexagonal phase. Previous investigators have used the symbols β' and γ for these phases, but it appears to the authors to be desirable to use a consistent notation throughout. The use of the symbol γ for a close-packed hexagonal structure appears very undesirable, since this symbol is usually employed for phases with a γ -brass structure.

4 W. Hume-Rothery, G. W. Mabbott, and K. M. Channel-Evans, *Phil. Trans. Roy. Soc.*, 1934 [A], **233**, 1.

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21 S. Ishida, H. Tajiri, and M. Karasawa, *Rep. Arcweld. Res. Inst. Tokyo Imp. Univ.*, 1937, **13**, 77.

22 I. Obinata and M. Hagiya, *Sci. Rep. Tōhoku Imp. Univ.*, 1936, [i], Honda Anniv. Vol., 715.

Sodium in Aluminium-Silicon Alloys

SODIUM can be added to aluminium-silicon alloys either by fluxing with sodium containing salts, by the direct addition of metallic sodium to the molten alloys, or by direct production of the alloy in the electrolytic cell under conditions in which sodium is simultaneously produced in appropriate amounts. These three methods are discussed in some detail in the June issue of *Aluminium Technique*.^{*} For the first method a mixture of sodium chloride and sodium fluoride, in the proportion of 1 to 2, may be used. The amount of salt is about 2% of the melt, and the temperature at which the melt is treated is 800° C. Stirring should be used to achieve more rapid reaction. If metallic sodium is used it should be wrapped in aluminium foil before being added to the alloy, which should be at a temperature of 750° C. The amount of sodium added should be about 0.05 to 0.1% by weight of the melt. If the sodium has been kept under paraffin oil, the coating formed in this solvent must be removed before introduction into the melt.

Both the above methods require a subsequent holding of the molten metal before pouring, to reduce the amount of sodium to the proper proportions, to achieve its uniform distribution, and to eliminate any gas which may have been introduced.

Production of sodium containing aluminium-silicon alloys by direct reduction has the advantage of a more uniform distribution of the sodium throughout the alloy.

Hard Facing to Resist Wear

Hard facing of new and worn parts is proving an easy and inexpensive means of keeping plant and machines in better serviceable condition than the normal base metal will permit. The most important economy its use effects is increased service life of the part treated. Brief reference is made to processes and materials applied by fusion welding.

APPRECIATION of the value of hard facing has greatly increased in recent years, largely as the result of the development of many processes of applying metals and alloys or treatments to surfaces which increase resistance to wear, corrosion, or both. Hard surfaces may be produced on metal parts of suitable composition by means of various forms of heat-treatment, such as carburising, nitriding, induction hardening, flame hardening, etc.; by metal spraying; by electro-deposition of chromium or other hard elements; by the fusion welding of hard alloys, or the cementing of special sintered carbides. In each section there has been considerable development, and although there is a great difference in the degree of hardness obtained by these methods, each has its own field of usefulness. It is not proposed to consider all these methods here, but to deal, somewhat briefly, with those processes which give relatively thick deposits and employ principally the oxy-acetylene and electric-arc methods of application.

It is generally recognised that the main purpose of hard facing is to reduce the wear of a particular working part or piece of a machine that is subjected to heavier service than the remainder of the machine, and by prolonging the serviceable life of this part the machine can remain in operation for a longer period, with a substantial reduction in maintenance and idle time costs. Generally, however, in addition to lower service costs, hard facing increases the life of the particular part, has a low coefficient of friction, is resistant to corrosion, and can be applied successfully to cheaper base metals. Hard-faced parts will outwear the normal base materials used from 2 to 25 times, depending upon the type of hard-facing alloy used and the service to which the part is subjected.

It is a well-known fact that toughness and hardness of metals do not go hand in hand. Many tools, implements, and machinery parts must be strong and tough because of their very design or function. If such parts are made of tool steel and are heat-treated, a compromise between hardness and toughness must generally be accepted. With parts designed for hard facing, however, the body or parent metal can be a steel that is expressly suited for strength requirements and shock resistance. Only the strength and cheapness of the bulkier base metal need be considered. Then the wearing surfaces can be hard-faced with a suitable wear-resistant material, thus creating a composite part which possesses both strength and hardness, each used where it is needed most. Often plain medium-carbon and low-alloy steels can be used as the base of the part. These have the necessary strength and shock resistance.

High quality materials are required for hard facing, but no one material can be applied for every purpose. Tungsten carbide may be the most suitable for use where friction is the deciding factor, providing little or no impact is encountered. For smooth bearing surfaces or corrosion resistance, the cobalt-base non-ferrous alloys are best. An austenitic alloy is used to withstand great shock. Zeyen¹ gives five groups of hard-facing materials, which are, in the order of increasing wear resistance, alloy steels containing up to approximately 20% of alloy constituents; iron-base alloys containing over 20% of alloying constituents; non-ferrous cobalt-chromium-tungsten alloys; crushed tungsten carbide, which, after welding, is embedded in a steel binder; and cast tungsten carbide inserts. The order of their toughness is approximately the opposite of wear resistance.

The hard-facing material in the first group generally includes manganese as an alloying element in addition to chromium. It has a relatively low melting point, good fluidity, favourable coefficient of expansion, and definite resistance to acid and alkaline liquids. It has a comparatively low cost. The structure consists of chromium carbides in an iron-manganese matrix. Rapid cooling results in tough austenitic deposits, whereas slow cooling produces hard deposits.

The second group comprises important cast alloys which contain little or no iron. These are the cobalt-chromium-tungsten alloys, which may also contain vanadium or molybdenum. Stillite is in this group as well as several other proprietary alloys. All are characterised by high resistance to acids and alkalis, low coefficient of friction, and high hardness up to 700° C. The melting point of these alloys is about 1,300° C., while the hardness varies from 40 to 60 Rockwell C. according to composition. The structure of these alloys consists of needles of chromium and tungsten carbides in a dendritic matrix.

The hard metals of the third, fourth and fifth groups consist mainly of carbides having high melting points, mainly tungsten carbide, and the rods used to apply them to surfaces are usually sintered, though in America they are cast. The hard metals in the third group are generally contained in steel tubes. Gas welding may be used or covered electrodes connected to the positive pole for electric welding. The particles of hard metals dissolve partially in the liquid steel from the outer tube. The deposit consists of an iron-tungsten matrix through which fine particles of tungsten carbide are disseminated.

In the fourth group powdered or granular tungsten carbide in its purest form is used to yield the highest hardness. The melting point is about 2,800° C. and the materials can be applied only by means of the carbon arc. The surface to be hard faced is strewn with the powder and is connected with the negative pole. The best results are obtained by welding several layers on to the steel base. The deposits are not held in a matrix of tough iron-tungsten alloy, as in the third group, and therefore can only be used in the absence of shock. In some cases the tungsten carbide powder is mixed with a suitable binder to facilitate application. The fifth group consists of small individual shapes or inserts of tungsten carbide which are bonded in place by means of a steel welding rod.

The most significant development in these alloys has been the increase in toughness of the harder classes of materials. Proprietary alloys and metals are available for the materials in practically all these groups. Many manufacturers apply these hard-facing materials to new parts before they are marketed. Hard-facing alloys are also being used where wear resistance alone is not the deciding factor, but where resistance to galling and pitting and high temperature corrosion is desired.

According to LeVan² most steels can be readily hard faced, but those containing 0.50% carbon and over usually must be heat-treated, and the hard facing of high-speed steel is unsatisfactory. Low alloy and stainless steels, grey cast iron, alloy cast iron, and nickel base alloys are also hard faced successfully. Brass, bronze, copper and aluminium alloys are difficult to hard face, because of their relatively low melting points.

All loose scale, dirt, rust, or other foreign substances should be removed from the areas to be hard faced, pre-

¹ Zeyen: "The Use of Surface Welding for Repair and Fabrication," *Autogene Metallbearbeitung*, 32, pp. 117-123, 133-141.

² LeVan: Contribution by the Machine Shop Practice Division of the American Society of Mechanical Engineers presented at June 17-20, 1940, Annual Meeting. *Mech. Eng.*, 62, pp. 459-464.

ferably by grinding or machining. The base metal should be clean. Where the hard-facing material is to be applied to a groove, corner, or recess, it is preferable to have all corners well rounded, otherwise local overheating of the base metal may result. If but a small section of a part is to be hard faced, a suitable shallow recess, groove, or indentation to receive the hard-facing metal may be prepared either mechanically or with the oxy-acetylene cutting blowpipe.

The same general rules apply to preheating before hard facing as apply to most welding operations. Parts are always preheated in cases where the application of the hard-facing alloy would otherwise cause them to warp or check, or where time and welding gases can be saved by preheating. Steels having over 0.50% carbon should usually be preheated. The larger the surface to be hard faced the higher the preheating temperature must be to avoid cracks. It is recommended that in nearly all circumstances the base metals should be preheated from 200° to 800° C., depending on size and sensitivity to hardening. Usually, preheating in a furnace is best, and, after hard facing, it is important to cool slowly after reheating to 700°-800° C.

Hard-facing metals and alloys are applied by either electric or oxy-acetylene welding processes, though non-ferrous alloys are best applied by means of the oxy-acetylene process. Generally, the oxy-acetylene flame allows close control over the operation and produces a smooth deposit. Particles of scale and foreign matter are easily eliminated by this method, and edges and corners can be readily formed. Other advantages include the elimination of any loss of expensive alloys by vaporisation or spattering, close control of the oxidising or reducing conditions of the atmosphere surrounding the molten metal, and accurate control over the degree of penetration of hard-facing alloy in the base metal. This latter advantage is of importance, especially with respect to the application of certain classes of hard-facing materials, some of which should be puddled in, while others are flowed on to the base material at a "sweating" heat. The oxy-acetylene process offers the facilities for producing either condition at the will of the operator.

The practice adopted with the oxy-acetylene flame varies with the nature of the metal being deposited. In the practice of depositing from the cobalt-chromium-tungsten hard-facing rods, for instance, the blowpipe should be adjusted to a carburising flame with the tip of the outer cone extending almost three times the length of the inner cone. A tip sufficiently large to hold the heat in the body metal should be used. The flame lends a little carbon to the hot, unmelted steel, converting a very thin surface film to a high-carbon steel with a much lower melting point. Hence this film will melt and the surface appear to "sweat" before any measurable part of the base metal is heated to its real melting range. When this occurs, the rod should be held directly beneath or beside the flame, and as the alloy melts from the rod it will spread freely over the sweating surface. Any scale met with or formed in the course of the work should be floated off by manipulation of the flame to avoid pinholes in the deposit.

The very thin original sweated layer forming the bond between the base metal and the hard-facing alloy possesses greater strength and toughness than either the base metal or the hard-facing alloy, due to the interalloying of the two metals in this thin layer.

With the metallic arc, a certain amount of interalloying is unavoidable, although it can be kept at a minimum through careful control. For covering of large areas the arc process is more economical, and it is frequently employed if rough surfaces and some cracking or checking of the hard-facing metal are not objectionable. Reversed polarity should be the rule in depositing all high-chromium alloys with the arc. Power consumption is fairly heavy.

The metallic arc is always used when hard-facing Hadfield's or 13% manganese steel, as, for instance, on

steam-shovel bucket teeth or crusher jaws and mantles. When this steel is heated to the welding temperature and cooled in the ordinary way, it becomes brittle. Furthermore, its coefficient of expansion is about 50% greater than that of a hard-facing alloy of the cobalt-chromium-tungsten type. The steel can be restored to its original ductility by quenching the hard-faced part from about 960° C. in water.

Vickers' Interim Dividend

At a meeting of the Board of Vickers Limited, held recently the following interim dividends were declared: $2\frac{1}{2}\%$ actual, less income-tax, on the preferred 5% stock; $2\frac{1}{2}\%$ actual, less income-tax, on the 5% preference stock; $2\frac{1}{2}\%$ actual, free of income-tax up to 6s. in the £ on the cumulative preference stock. Payment will be made on August 23, 1940.

Awards of the North-East Coast Institution of Engineers and Shipbuilders

THE Council of the above Institution has awarded the Institution Scholarship, valued at £100, to Ian Welsh Goodlet; a Bursary of £50 to Eric William Simpson; the M. C. James' Medal to Mr. H. J. Tapsell, of the National Physical Laboratory, for his paper on "Creep at High Temperatures"; the Thomas Fenwick Reed Medal to Mr. Harry Chilton; and the R. L. Weighton Medal to Wilfred Bailey.

International Nickel Company's Casting Research Facilities

AN experimental foundry and mill has recently been installed as an adjunct to the International Nickel Company's research laboratory at Bayonne, N.J., which is somewhat unique in the elaborateness of its equipment. For the production of alloy cast irons there is now available a cupola having a melting rate of 1.6 tons per hour, a direct-arc furnace of 1,000 lb. capacity, and an indirect-arc furnace of 350 lb. size, in addition to several smaller crucible and induction furnace melting units. Thus, by increasing the maximum unit of production from 30 lb. to 1,000 lb., a single heat can turn out enough metal of uniform composition to permit exhaustive programmes of research on cast iron, free from uncertainty due to unavoidable variations between heats which formerly occurred when it was necessary to make several runs to complete a study.

Accuracy and close control to eliminate variables to the maximum extent, are primary considerations in research work, and the experimental foundry has been equipped with practically every known means to serve these ends. Moulding sand is tested for control of moisture content, crushing strength and permeability to air or gas. The cupola blast is air-conditioned, may be preheated, and its humidity regulated by specially-designed equipment, enabling uniform melting conditions to be obtained. The melting and pouring temperatures of every heat are carefully checked by optical pyrometers. Not only can the desired chemical compositions be obtained to closer limits with this careful control, but the physical properties of the metals produced are more homogeneous and the results are more dependable.

All experimental samples are carefully classified as to composition, production procedure, mechanical test results, etc. Samples of each heat, in addition to those used for chemical, physical, and corrosion or other special tests, are filed, so that records of our test may be checked against subsequent series of tests under varying conditions.

The increased production capacity provided by the new facilities make it possible for the laboratory staff to undertake the study of many metallurgical problems formerly closed to investigation, and to carry out experiments with greatly increased speed and effectiveness.

Recent Developments in Materials, Tools and Equipment

New Experimental Tinning Machine

MANY of the discontinuities in the coating of ordinary tinplate are traceable to ridge and valley structures in the tin layer, formed in the tinning machine. With the object of producing perfectly smooth tin coatings, an experimental tinning unit, designed by members of the staff of the International Tin Research and Development Council, has been installed in the laboratories at Greenford, significant reductions in porosity have already been secured, although the machine is only in its preliminary stages of development.

Transverse Ridges

Transverse ridge and valley structures in the coating may result from mechanical imperfections in the drive and loading of the grease-pot rollers, and from vibration of the sheet as it passes through the grease-air interface. In this experimental machine, Fig. 1, the drive is arranged as follows: Separate worm gears (a) are connected by flexible couplings (b) to each roller neck. Sideways movement of one of each pair of rollers is provided for by spring stuffing boxes (c). By this method of drive two important advantages are secured:

(a) In common practice the sideways movement of the rollers is obtained by working the driving pinions on incorrect-pitch circle diameters. The use of flexible couplings specially designed to give even torque avoids this undesirable feature and eliminates the typical irregular movement of the rollers.

(b) Since the driving mechanism is not required to work in hot palm oil, it is possible to make it a precision engineering job.

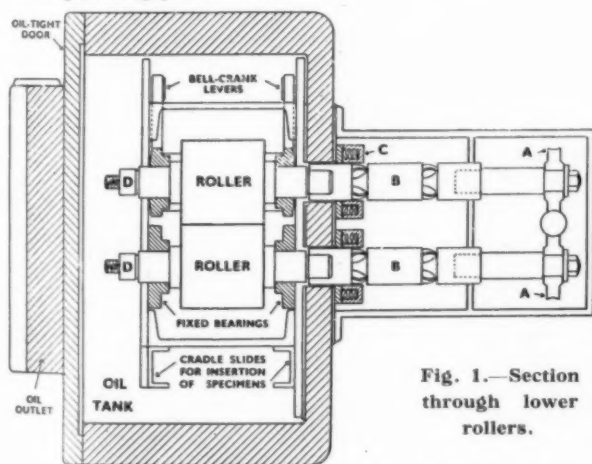


Fig. 1.—Section through lower rollers.

The rollers themselves are loaded through bell-crank levers of heavy construction integral with the neck bearings of the rollers. To compare this type of drive with the usual type, ordinary driving pinions may be slipped on at (d).

In Fig. 2, the cast-iron bell-crank levers (1) are pivoted at (2), and the loading device is connected at (3). Upward pull at (3) provides increased pressure between the rollers. (4) is the neck bearing of the movable roller of the pair (4), (5).

Longitudinal Ridges

The formation of longitudinal ridge and valley structures has been shown to be inevitable in any physical process involving the distribution of a layer of liquid on a solid surface by means of a roller, unless the conditions are such that the roller is not wetted by the liquid.

In the manufacture of tinplate the tin ridges are formed at the tin-grease interface in the nip of tin lying beneath the grease at the top of the rollers in the grease-pot. In the experimental machine the upper pair of rollers are specially coated, so that they are not wetted by molten tin. The resulting alteration in the shape of the tin meniscus is represented diagrammatically in Fig. 3.

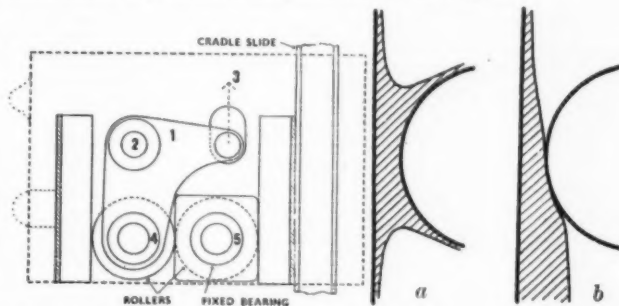


Fig. 2.—Front view of upper roller frame.

Fig. 3. (a) Roller wetted by molten tin; (b) Non-tinning roller.

Encouraging Results

The main features of the new experimental tinning machine may, therefore, be summarised as the provision of perfectly smooth drive and the replacement of the top-most pair of rollers in the grease-pot by a pair of non-tinning rollers. The results already obtained indicate that by these means it is possible to obtain tin coatings having only from 50 to 75% of the number of discontinuities usually associated with commercial coatings of similar thickness.

A Storage Refrigerator for Aluminium Alloy Rivets

THE demand for increased aircraft production has greatly increased the use of high-strength aluminium alloys of the duralumin type. These alloys develop improved physical properties as a result of age-hardening following a normalising solution treatment, which confers upon them the desired high strength-to-weight ratio. The improvement in physical properties resulting from age-hardening normally takes place at room temperature over a period of about four days, but the change in ductility becomes apparent within an hour or so of solution treatment, thus, when the material requires to be bent, stretched or subjected to deformation, the operation must be carried out within a comparatively short time of solution treatment to avoid rupture of the material. The reduction of ductility does not affect appreciably the usefulness of the finished article, but it constitutes a hindrance to its fabrication.

When the reduction in ductility is such that it would be impossible to carry out the required work on the material without cracking it has been customary to reheat-treat the material to restore the ductility. Although the operation is relatively simple and can be repeated many times without detriment to the properties of the material, it constitutes a real hindrance to speed in production. Considerable trouble was encountered with rivets, for instance, and it was found necessary to institute quite elaborate systems in aircraft and other factories where these were used in large quantities, to remove from the shops all rivets which had been heat-treated more than two hours beforehand, and to issue in their place a freshly solution-treated supply.

It has long been known that age-hardening could be retarded and the necessary degree of ductility retained by storing the solution-treated material at low temperature, but it is only in recent years that the possibilities of low-temperature storage has been seriously considered. It was found that, when stored in a refrigerator operating at a temperature of -10 to -15°F. , alloys of the duralumin type, which had been given a solution treatment, could be kept for much longer periods without appreciable loss of ductility, thus require heat-treatment at less frequent intervals. Refrigerators for this purpose were first applied for storing rivets, and among manufacturers of equipment for this purpose, Coldair Ltd., an associated company of the General Electric Company Ltd., has been active. They have made several notable installations, including one at the Gloster Aircraft Factory.

The refrigerator cabinets which are used for retarding the age-hardening of rivets in this factory are designed to

facilitate rapid loading and unloading of the low-temperature compartment with the minimum rise of temperature arrangements having been made to maintain an internal temperature of -10° to -15°F. , with an ambient temperature of 80°F. They are constructed from heavy-gauge material which ensures them a long and useful life under hard factory conditions. The capacity is 2.9 cubic feet, and the galvanised steel storage compartment which measures 22 in. \times 10 $\frac{1}{2}$ in. \times 21 $\frac{1}{2}$ in. deep, holds 500 lb. of rivets at a filling. Overall dimensions are 4 ft. 4 in. long \times 2 ft. 1 in. wide \times 2 ft. 8 in. high, and the general finish white or black enamel with silver-finished steel frame.

The specification includes 4-in. compressed-cork insulation, galvanised mild steel working top, thermostatic temperature control, and dial thermometer. The condensing unit comprises a standard $\frac{1}{2}$ -h.p. twin-cylinder compressor and $\frac{1}{2}$ -in. solid-drawn copper cooling coils.

Corrosion-Resistance of Zinc Alloys

By Louis Light, M.Sc.

The results are given of some researches on a number of zinc alloys of which the principal alloying element is aluminium, and the effect of small additions of other elements is discussed.

ZINC die-casting alloys have been in use for a good many years. Some aspects of their properties as influenced by various impurities and alloying metals are discussed in *La Metallurgia Italiana*, of April, 1940, by Piontelli and Cremascoli, who describe some results of their own researches and put forward a theory of intercrystalline corrosion in presence of lead.

The chief alloy metal used with zinc is aluminium, the limits of which are specified as approximately 3.5 to 4.5% (A.S.T.M. Specification No. 23 stipulates 3.5 to 4.3% Al for a die-casting alloy). In this proportion the mechanical working properties of zinc are greatly improved, fine grain structure is developed and thereby opposes the characteristic zinc crystallisation. As little as 0.05 to 0.5% aluminium minimises the solubility of iron in zinc, and so reduces contamination with the former metal during die-casting work. Zinc-aluminium alloys can be aged by keeping for three to four hours at 100°C. in a dry atmosphere, after which no change in dimensions or mechanical properties need be feared as long as conditions favouring intercrystalline corrosion are absent.

In the latter connection a profound influence is exercised by the lead present in traces in all but the very purest grades of commercial zinc. The practically complete insolubility in the solid state of lead in zinc results in separation of the lead as a distinct phase (generally as globules) in the intercrystalline cement. This structure is not objectionable in zinc alone. In zinc-aluminium alloys, however, ageing subsequently leads to separation of aluminium or crystals rich in aluminium. It is the electrochemical system arising from contiguity of this aluminium-rich phase to the lead which makes the former so liable to attack, and thus accounts, in the author's view, for the severity of the intercrystalline corrosion noted under these conditions in practice. This effect is especially marked in hot, moisture-laden air (e.g., in the tropics). The favourable effect on this type of corrosion resulting from the presence of certain other alloy metals like magnesium and copper may be attributed to the formation of mixed crystals with different electrochemical reactions.

Apart from the reduced intercrystalline corrosion brought about by a trace of copper in zinc-aluminium alloys, this metal also improves the tensile strength, hardness and weldability. Magnesium, likewise, compensates up to a

point for the harmful effect of lead. Nevertheless, the increased hardness resulting from the separation of distinct Mg-Zn compounds is offset by increased brittleness. Another drawback is the tendency of the castings to develop cavities after cooling. Magnesium is, nevertheless, a standard ingredient of die-casting alloys in the United States (e.g., in the proportion of 0.03 to 0.8%).

Experiments were carried out on the behaviour of zinc-aluminium alloys containing a small amount of lithium. This metal forms compounds such as LiZn_2 and Li_2Zn_3 at higher temperatures, although practically insoluble in zinc at the ordinary temperature. It promotes fine grain structure, increases tensile strength and hardness. Like magnesium, it is a corrective to the presence of lead, but does not introduce the same risk of cavitation. Drawbacks to its general adoption are the losses consequent upon its ease of oxidation and the high price.

It is usual to specify an extremely low limit for cadmium in zinc-aluminium alloys (e.g., 0.005%), but the authors were unable to reach a definite conclusion as to its influence. Tin is similarly kept very low as it aggravates intercrystalline corrosion and interferes with the plastic working qualities, although it promotes fine grain.

An examination was made of the influence of nickel. The diagram of state Zn-Ni is similar to that of Zn-Cu. Although nickel has a favourable effect on the tensile strength and steam-resistance of zinc alloys, it reduces the resistance to more drastic corrosive factors. It is more useful in copper-free zinc alloys.

The authors also discussed the factors making for production of satisfactory castings by the die-casting process. Overheating during manufacture of the alloys at all stages is to be avoided owing to inevitable oxidation of zinc and aluminium, and the probability of increased occlusion of gas. Elimination of oxides and gases from the fused mass is by no means an easy matter. It was found advisable to melt the alloys at 385° to 390°C. , and to operate at a casting temperature of 395° to 410°C. An excessive temperature during die casting is likewise accompanied by oxide formation and cavitation. The effects of overheating cannot be rectified by remelting. A too-low casting temperature is again undesirable, because embrittlement results from the presence of large preformed crystals.

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